TRANSACTIONS OF THE AMERICAN • SOCIETY • FOR METALS

DECEMBER, 1934

Volume XXII Number 10

The TRANSACTIONS *of the*

AMERICAN SOCIETY FOR METALS

formerly

AMERICAN SOCIETY FOR STEEL TREATING

Published quarterly and Copyrighted, 1934, by the AMERICAN SOCIETY FOR METALS
7016 Euclid Avenue, Cleveland, Ohio

SUBSCRIPTIONS: (members) $2.50 per year
(non-members) $5.00 per year, $2.00 per copy
Foreign (non-members) $6.50 per year, $2.50 per copy

Entered as second class matter, November 9, 1931, at the Post Office at Cleveland, Ohio, under the Act of March 3, 1879

RAY T. BAYLESS, *Editor*

Vol. XXII | December, 1934 | No. 10

The object of the Society shall be to promote the arts and sciences connected with either the manufacture or treatment of metals, or both.

Officers and Board of Trustees

B. F. SHEPHERD, President
Ingersoll-Rand Co., Phillipsburg, N. J.

R. S. ARCHER, Vice-President
Republic Steel Corp., Chicago

EMIL GATHMANN, Treasurer
Gathmann Engineering Co., Baltimore, Md.

W. H. EISENMAN, Secretary
7016 Euclid Ave., Cleveland

TRUSTEES

W. H. PHILLIPS, Past President
Molybdenum Corp. of America, Pittsburgh

WALTER MATHESIUS
Illinois Steel Co.
Chicago

GEORGE B. WATERHOUSE
Massachusetts Institute of Technology
Boston

W. P. WOODSIDE
Climax Molybdenum Co.
Detroit

E. C. BAIN
United States Steel Corp.
Kearny, N. J.

Publication Committee

G. R. BROPHY, Chairman — Schenectady, N. Y.
RAY T. BAYLESS, Secretary — 7016 Euclid Ave., Cleveland

Members:

W. L. Cockrell, Tri-City '36
M. H. Frommann, Worcester '35
R. L. Kenyon, Cincinnati '35
R. F. Mehl, Pittsburgh '36
C. B. Sadtler, Rockford '35
Claud L. Clark, Detroit '36
A. E. Focke, Indianapolis '35
T. L. Joseph, North-West '35
Joseph Winlock, Philadelphia '37
E. S. Davenport, New Jersey '37
D. F. McFarland, Penn State '37
W. P. Sykes, Cleveland '35
H. S. Taylor, Golden Gate '35

Grain-Size Symposium

HELD DURING THE SIXTEENTH ANNUAL CONVENTION
OF THE SOCIETY IN NEW YORK CITY THE WEEK OF
OCTOBER 1, 1934

Table of Contents

Grain-Size in Metals, with Special Reference to Grain Growth in Austenite—By M. A. Grossmann 861
Discussion 877

General Relations Between Grain-Size and Hardenability and the Normality of Steels—By E. S. Davenport and E. C. Bain 879
Discussion 921

Grain-Size in Relation to Machinability and Other Properties of Bessemer Screw Steel—By H. W. Graham 926
Discussion 940

Grain-Size Control of Open-Hearth Carbon Steels—By S. Epstein, J. H. Nead and T. S. Washburn 942
Discussion 973

The P-F Characteristic of Steel—By B. F. Shepherd 979
Discussion 1001

Effect of McQuaid-Ehn Grain-Size on Hardness and Toughness of Automotive Steels—By H. W. McQuaid 1017
Discussion 1036

Effects of McQuaid-Ehn Grain-Size on the Structure and Properties of Steel—By Philip Schane, Jr. 1038

The Effect of McQuaid-Ehn Grain-Size of Steel in Forging—By W. E. Sanders 1051
Discussion 1067

Influence of Grain-Size on the High Temperature Characteristics of Ferrous and Nonferrous Alloys—By A. E. White and C. L. Clark 1069
Discussion 1089

The Importance of Grain-Size of Sheet Steel for Deep Drawing—By Reid L. Kenyon 1099
Discussion 1115

The Influence of Grain-Size on Magnetic Properties—By W. E. Ruder 1120
Discussion 1132

Factors Determining the Impact Resistance of Hardened Carbon Steels—Howard Scott 1142
Discussion 1164

GRAIN-SIZE IN METALS, WITH SPECIAL REFERENCE TO GRAIN GROWTH IN AUSTENITE

By M. A. Grossmann

Abstract

Since steel may exist either as austenite or as ferrite and carbide (depending on the temperature), the grain sizes of these two structures must obviously be considered separately.

Ferrite grain-size is influenced by heat treatment, particularly after prior cold work, and the laws of variation of ferrite grain size are rather well understood.

The laws of variation of austenite grain-size still call for much study. Data are presented here for some lower carbon steels, showing how the austenite grain-size developed at any temperature may be affected profoundly by prior treatment of the steel. Prior heat treatment and prior hot work may shift the coarsening temperature of austenite markedly. In this connection, it is pointed out how prior cold work, prior hot work and prior heat treatment may affect the "McQuaid-Ehn grain-size."

IN APPROACHING a symposium on grain-size in metals, it seems appropriate first to define the precise nature of the problem, and then if possible to add a few data to our present knowledge of the subject.

Crystal Structure in Metals

Almost as soon as the microscope was directed at metallic bodies, it was found that they consisted of minute grains. It was not until 1898, however, that Heyn (1)* demonstrated after a study of etching pits that these grains were in reality crystals, a fact not previously recognized because of the irregular outlines of the grains. The name "crystals" had until then been applied only to mineral crystals, with their symmetrical form and regular cleavage. When Laue began the study of crystal structures with X-rays in 1912, the fact that metals were crystalline became a familiar circumstance. It is now known

*The figures appearing in parentheses refer to the bibliography appended to this paper.

A paper presented as part of the Grain-Size Symposium of the Sixteenth Annual Convention of the Society held in New York City the week of October 1, 1934. The author, M. A. Grossmann, a member of the Society, is research engineer, Illinois Steel Company, Chicago. Manuscript received August 2, 1934.

further that even those properties of crystals which are associated with minerals (regularity of outline and manner of cleavage) may be observed also in metals. Examples of large crystals of steel grown in shrinkage cavities of steel ingots are well known. Cleavage was demonstrated among others by Edwards and Pfeil (2), as shown in Fig. 1, which shows the cleavage in large single crystals of iron, obtained by striking the crystals a sharp blow. Regularity of form may on occasion be observed also in microstructures. Thus Fig. 2

Fig. 1—Cleavage in Large Single Crystals of Iron Obtained by Striking the Crystals a Sharp Blow. (After Edwards and Pfeil)

Fig. 2—Photomicrograph of a Crystal Grain of Austenite of Symmetrical Shape Formed within a Single Grain of Ferrite in 27 Per Cent Chromium Steel Heated Several Hours at 2300 Deg. Fahr. × 500.

shows a crystal grain of austenite of symmetrical shape. It formed within a single grain of ferrite, in 27 per cent chromium steel, when the steel was heated for several hours at 2300 degrees Fahr. (1260 degrees Cent.).

Such regularity of outline as shown in Fig. 2 is however unusual, since it is a result of unobstructed growth, whereas ordinarily there is mutual interference during growth, with its resultant familiar irregularity of form. The absence of crystal outline, however, does not affect the inner crystal structure.

One important manifestation of this crystallinity is the manner of plastic flow. Crystals of metal can flow or deform more easily in some directions than in others (3). Since the individual crystals

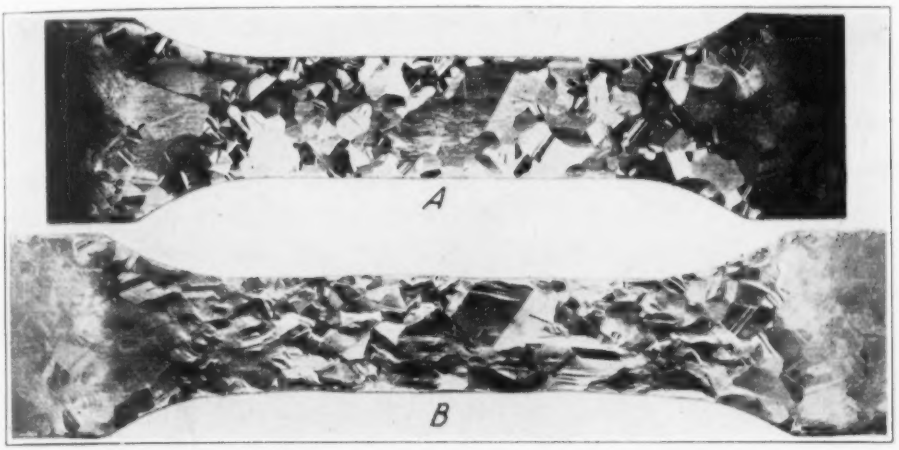


Fig. 3A—Test Piece of 18-8 Stainless Steel Before Stressing in Testing Machine.
Fig. 3B—Same Test Piece After Being Elongated 27 Per Cent in 2 Inches.
Specimens Etched to Show Individual Grains.

in a piece of metal have different orientations, it follows that when a piece of metal is deformed, each of its individual grains will deform in its own particular manner. The manner of deformation is influenced by the orientation of the crystal, the direction of the applied stress and the obstructing action of adjacent grains. These individual manners of deformation may be observed readily in a coarse-grained piece of metal, such as the 18 per cent chromium-8 per cent nickel steel of Fig. 3. The coarse grains were made by heating the steel for 24 hours at about 2400 degrees Fahr. (1315 degrees Cent.), whereupon a tensile test piece was cut from the block. Fig. 3A shows the test piece, etched to show the individual grains. Fig. 3B shows the same test piece after being elongated in a tensile machine, the elongation being 27 per cent in 2 inches. It will be seen that the individual grains have deformed, each in its own manner.

Relation Between Actual Grain-Size and Other Properties

Considerations such as these have led to many investigations of

the relation between the crystal size and the accompanying mechanical properties. A few references to previous work (4), (5), (6), will be found in the appended bibliography, and new data are to be presented in the present symposium.

The mechanical properties at room temperature, however, do not by any means constitute the only properties which have been correlated with grain-size. There are other properties which have been studied at room temperature, such as magnetic permeability (7) and watt loss in sheets for electric purposes (8). There have also been studies of properties at elevated temperatures, such as tensile strength and creep strength.

Change of Actual Grain-Size

Since it has been observed that there are relationships between the grain-size of a piece of metal and certain of its properties, and since the size of the grains may be changed by heat treatment, it becomes important to inquire into the laws of change of grain-size.

In one field of grain growth, a very considerable amount of work has already been done, namely the grain growth which takes place when metal is cold-worked and then heated. These studies had their inception in the work of Sauveur (9), who showed that a piece of steel, deformed under a Brinell impression and then reheated, showed a wide range of grain growth tendencies. In particular, it was found that a certain moderate amount of cold work followed by heating in a certain range of temperature, led to the formation of abnormally large grains, and that a whole range of grain-sizes was produced by different amounts of cold work and different reheating temperatures. This behavior, which was found to be general in metals, has been studied in detail in iron by Chappell (10), in alpha brass by Mathewson and Phillips (11), in aluminum by Carpenter and Elam (12), and in other metals and alloys.

Relation Between Austenite Grain-Size and Other Properties

Now the relationships which have been mentioned up to this point may properly be called grain-size relationships—that is to say, they refer to a relationship between the properties of the piece of metal being tested and the size of the crystals which compose the piece. In an entirely different category are the studies of austenite grain-size in steel, as exemplified in the widely used McQuaid-Ehn

test. Here the steel is examined not for its actual grain-size while being tested but to discover what its grain-size had been previously, namely when it consisted of crystals of austenite, at high temperature. This point can scarcely be emphasized too strongly. The McQuaid-Ehn grain-size, or any other record of austenite grain-size, is a record of the size of the austenite crystals which existed at high temperature. The size of these austenite grains in any individual piece of steel may be varied in a number of different ways, and so may have little or nothing to do with the size of the grains in the piece after it has cooled and transformed to ferrite. This is the pitfall in attempts to correlate austenite grain-size with other properties of steel.

But when properly interpreted, austenite grain-size furnishes extremely important clues to certain behaviors. McQuaid and Ehn (13), in their original work on this subject, showed that when steels are carburized and the austenite grains in the case are coarse after carburizing, the steel hardens more readily in quenching than when the grains are fine. This behavior was eventually explained clearly and beautifully by Bain (14) as a result of his studies of hardenability.

Change of Austenite Grain-Size

Now it may be said that there is rather good understanding of the manner in which steel (particularly low carbon steel) undergoes changes in actual grain-size when it is cold worked and then heated.

Less well understood are the ways in which austenite grains in steel grow when they are heated. Austenite is the stable structure in the range of temperature from about 1300 to about 2500 degrees Fahr. (705 to 1370 degrees Cent.), and at these high temperatures it is readily imaginable that grain growth might take place. But the temperature at which grain growth begins varies with the steel, so that numerous tests must be carried out in order to learn the austenite characteristics of a piece of steel. The carburizing test for austenite grain-size, the McQuaid-Ehn test (13), was shown to set forth very clearly the austenite grain-size under one set of conditions, and this grain-size was correlated with other properties of the steel. Unfortunately, the assumption was subsequently made by other workers that the austenite grain-size so discovered was characteristic of that piece of steel, regardless of any heat treatment of the piece before carburizing. It was also assumed that hot working (forging, rolling, etc.) was without influence on the McQuaid-Ehn

grain-size. It was furthermore assumed that the grain-size in the carburized case was the same as in the uncarburized core. It has already been demonstrated that some of these assumptions were unwarranted, and some evidence will be offered here which indicates that the remainder are likewise inconsistent with the facts.

Variations in Austenite Grain-Size

In considering variations in austenite grain-size, it is necessary to bear in mind the manner in which austenite grains form and the way in which they grow. This mechanism, already described elsewhere (15), may be summarized briefly as follows: When, upon heating, the steel reaches the temperature at which it begins to transform to austenite, the first very minute austenite crystals form within a pearlite island or in the boundaries between ferrite grains. These small austenite grains then grow across the pearlite islands and across the ferrite grains, until they meet other austenite grains which have grown from other nuclei. When all the austenite grains have thus met, so that the whole piece consists of austenite, it is clear that a certain set of grain-sizes will have been established.

Table I
Composition of Steels Investigated

	Per Cent							
Steel No.	C	Mn	P	S	Si	Ni	Cr	Mo
1	0.17	0.79	0.009	0.014	0.29	0.14	0.06	....
2	0.17	0.51	0.007	0.004	0.22	1.68	0.07	0.23
3	0.33	0.67	0.019	0.016	0.24	1.17	0.59	....
4	0.18	0.48	0.010	0.006	0.25	1.73	0.03	0.24
5	0.16	0.47	0.009	0.005	0.25	1.70	0.04	0.23

It should be mentioned that this initial austenite grain-size is not appreciably affected by the rate at which the steel passes through the transformation temperature range. The following experiment illustrates this point. Steel Number 1 of Table I, a low carbon plain carbon steel, was heated through the transformation range at several rates, as follows.

1. *Slow heating.* The piece was packed in a box of carburizer, the box placed in a furnace which was at somewhat below 800 degrees Fahr. (425 degrees Cent.), and the furnace then brought to 1700 degrees Fahr. (925 degrees Cent.). The time from room temperature to 1700 degrees Fahr. (925 degrees Cent.) was about 2 hours and 30 minutes.
2. *Moderate Rate of Heating.* The piece was packed in a box

of carburizer, and the box placed in a furnace which was already at 1700 degrees Fahr. (925 degrees Cent.). The time from room temperature to 1700 degrees Fahr. (925 degrees Cent.) was approximately 30 minutes.

3. *Rapid Heating.* A box of carburizer was heated to 1700 degrees Fahr. (925 degrees Cent.), whereupon the cold piece of steel was placed in the hot carburizer. The time from room temperature to 1700 degrees Fahr. (925 degrees Cent.) was estimated to be less than 5 minutes.

In all cases, the time of holding at 1700 degrees Fahr. (925 degrees Cent.) was about 2 hours, whereupon the boxes were cooled in air. The austenite grain-size which had formed was then determined in the usual manner by observing the carburized case. The three samples which had been given the three different rates of heating all showed the same austenite grain-size, about Number 5 on the A.S.T.M. chart, which had been found in other tests to be the size of austenite grains to which this steel transformed on slow heating.

These initial austenite grains have some stability, so that no grain growth takes place if the steel is heated for considerable lengths of time just above the transformation temperature, or even if it is heated to somewhat higher temperatures. At some still higher temperature, which is conveniently termed the coarsening temperature, some grains begin to grow at the expense of others, and after a time the whole piece consists of larger grains. This coarsening temperature varies greatly in different steels—it is common to find it anywhere in the range from 1500 to 2100 degrees Fahr. (815 to 1150 degrees Cent.). It is well to emphasize the fact that the grain-size does not change below the coarsening temperature:—If for example the coarsening temperature is 2100 degrees Fahr. (1150 degrees Cent.), then the steel may be heated many hours at 2000 degrees Fahr. (1095 degrees Cent.) and its grain-size will still be precisely the same as it was when it first transformed to austenite upon heating say at 1450 degrees Fahr. (790 degrees Cent.).

The austenite grain-size in any piece of steel, then, must be considered first in the light of the initial austenite grain-size which forms upon transformation and then the size of the grains which result from the coarsening of those initial grains. The initial austenite grains are usually in the range of size from Number 5 to Number 8 (A.S.T.M. chart) and the coarsened grains usually range in size from Number 1 to Number 4.

Now it is well known that different steels, though of the same composition as regards elements ordinarily determined, may have widely different coarsening temperatures. These differences are due to the presence of different amounts of grain-growth inhibitors.

But attention must now be directed to the fact that the coarsening temperature in a steel is not by any means constant for that piece of steel. It may be varied by prior heat treatment, by hot working

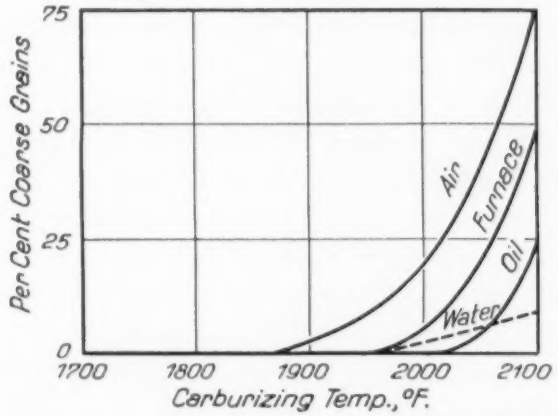


Fig. 4—Carburized S.A.E. 4615 Steel in which the Coarsening Temperature was Influenced by Prior Heat Treatment.

and probably by cold working. Fig. 4 shows an instance of S.A.E. 4615 nickel-molybdenum steel, in which the coarsening temperature was influenced by prior heat treatment. The steel, Number 2 in Table I, was treated as follows. Four pieces were placed in a furnace and heated to 1700 degrees Fahr. (925 degrees Cent.). One piece was quenched in water, one quenched in oil, one cooled in air and the remaining one cooled with the furnace. The only variable, then, was the rate of cooling from 1700 degrees Fahr. (925 degrees Cent.). The pieces were thereupon carburized for eight hours at a series of temperatures and the coarsening observed. The results are given in Fig. 4. As suggested elsewhere (15), a convenient method of plotting the coarsening is to indicate for each temperature of carburizing the percentage of the observed area which is occupied by coarse grains, that is to say, grains which have coarsened from the initial austenite grain-size which had been formed during transformation. Fig. 4 and the subsequent diagrams are plotted in this manner.

It will be observed that the sample cooled in air began to coarsen (in the subsequent carburizing) at a considerably lower temperature

than the other samples. Also, at higher carburizing temperatures, the air-cooled sample coarsened more than the others. Attention is directed to a peculiar behavior in the water-quenched sample, in that it did not coarsen as rapidly with increase in carburizing temperature as did the others. This behavior will be discussed again later.

The marked coarsening of air-cooled (normalized) samples has been observed in many steels. It is not at all uncommon to find the

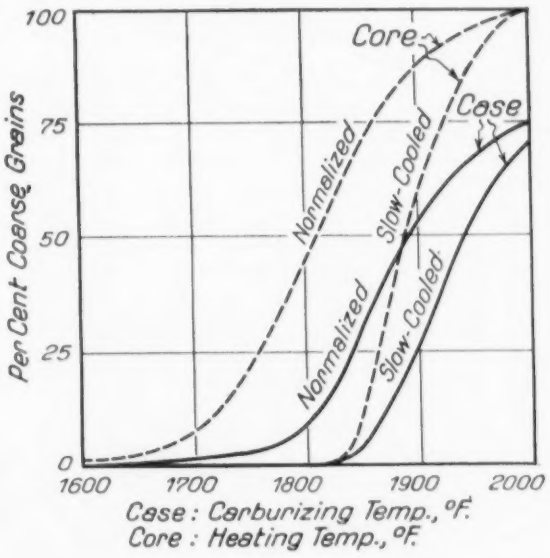


Fig. 5—Carburized S.A.E. 3135 Steel in which there is a Marked Lowering of the Coarsening Temperature Due to Normalizing Treatment.

coarsening temperature of an annealed low carbon alloy steel lowered as much as 150 degrees Fahr. by normalizing. A pronounced instance is shown in Fig. 5. Here Steel Number 3 of Table I, an S.A.E. 3135 steel, was available in the form of 6x6-inch billets, which had been slowly cooled from the rolling. A 3x3x6-inch piece was cut from such a billet, heated to 1600 degrees Fahr. (870 degrees Cent.) and cooled in air. Samples from this latter piece were then carburized at a series of temperatures and their grain coarsening compared with samples from the original slowly-cooled billet. The results assembled in Fig. 5 show a marked lowering of the coarsening temperature due to the normalizing treatment.

These observations raise a question as to a similar effect when the heat treatments are repeated. Steels 4 and 5 of Table I, both nickel-molybdenum carburizing steels similar to Steel Number 2,

were treated as follows. The steel was selected from 6x6-inch billets which had been slowly-cooled from the rolling.

B = Original billet.
N1 = Normalize from 1600 degrees Fahr. (870 degrees Cent.), once.
N3 = Normalize from 1600 degrees Fahr. (870 degrees Cent.), three successive treatments.
Q1 = Quench in water from 1600 degrees Fahr. (870 degrees Cent.), once.
Q3 = Quench in water from 1600 degrees Fahr. (870 degrees Cent.), three successive treatments.

After these treatments, the steels were carburized at a series of temperatures, and the results are assembled in Figs. 6 and 7. It will be seen that repeated treatments are even more effective in increasing the susceptibility to coarsening than are single treatments. Thus the pieces which had been normalized three times were in both cases (Figs. 6 and 7) almost completely coarsened in the 8 hours at 2000 degrees Fahr. (1095 degrees Cent.), whereas the original billet pieces were scarcely coarsened at all at this temperature.

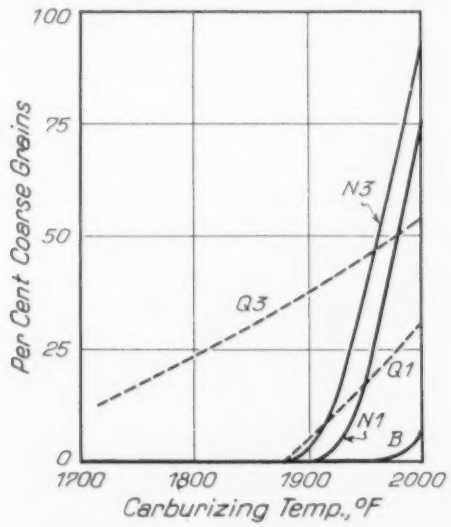


Fig. 6—Steel No. 4 (Table I) Carburized After Preliminary Heat Treatments Almost Completely Coarsened in 8 Hours at 2000 Degrees Cent., Whereas the Original Billet Pieces Were Scarcely Coarsened at all at this Temperature.

Attention is now directed again to the pieces which were quenched in water (Q1 and Q3). It was observed in Fig. 4 that the piece quenched in water, while it began to coarsen at a lower temperature than the piece quenched in oil, nevertheless did not display as much increase in coarsening when the carburizing temperature was

raised. This same behavior is displayed in the steels depicted in Figs. 6 and 7, particularly those which had been quenched in water three times. These pieces coarsened rather notably at temperatures as low as 1800 degrees Fahr. (980 degrees Cent.), yet even at 2000 degrees Fahr. (1095 degrees Cent.), the increase in coarsening was not very marked. This behavior may be contrasted with that of the pieces normalized three times—these latter barely showed signs of incipient

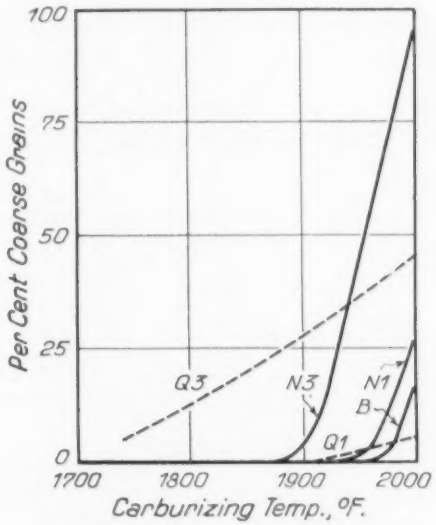


Fig. 7—Steel No. 5 (Table I) Carburized After Preliminary Heat Treatments Almost Completely Coarsened in 8 Hours at 2000 Degrees Cent., Whereas the Original Billet Pieces Were Scarcely Coarsened at all at this Temperature.

coarsening at 1900 degrees Fahr. (1040 degrees Cent.), and yet were almost completely coarse at 2000 degrees Fahr. (1095 degrees Cent.).

As to the reasons for this change in coarsening temperature due to heat treatment, it had been thought that it might be directly associated with differences in initial austenite grain-size. It will be recalled (15) that normalizing of low carbon alloy steels results in a smaller ferrite grain-size, and this in turn leads to a smaller initial austenite grain-size when the steel transforms to austenite in heating. Since the normalized steels, with their smaller initial austenite grain-size, have greater susceptibility to coarsening, it was thought that the greater surface energy in the small grains might be the sole reason for the change in coarsening temperatures. This, however, cannot be the case, as the following data will show.

The four samples of Fig. 4 were compared as to initial austenite

grain-size (i.e. before coarsening) and were found to be as follows:

Size of Initial Austenite	Manner of Cooling	A.S.T.M. Rating	Per Cent Coarsened at 2100 Degrees Fahr.
Coarsest	Furnace	6 to 7	50
Next finer	Water	7	10
Next finer	Oil	7 to 8	25
Finest	Air	7 to 8	75

There is thus no regular relationship between the initial austenite grain-size and the amount of coarsening at 2100 degrees Fahr. (1150 degrees Cent.).

A similar examination of the samples of Fig. 6 and 7 gave the following data.

Fig. 6, Steel Number 4

Size of Initial Austenite	Manner of Cooling	A.S.T.M. Rating of Initial Austenite	Per Cent Coarsened at 2000 Degrees Fahr.
Coarsest	Original billet	5 to 6	10
Next finer	Water, once	6 to 7	30
Next finer	Air, once	7 to 8	75
Next finer	Air, three times	7 to 8	95
Finest	Water, three times	7 to 8	55

Fig. 7, Steel Number 5

Size of Initial Austenite	Manner of Cooling	A.S.T.M. Rating of Initial Austenite	Per Cent Coarsened at 2000 Degrees Fahr.
Coarsest	Original billet	5 to 6	15
Next finer	Air, once	7	25
Next finer	Water, once	7	5
Next finer	Air, three times	7 to 8	95
Finest	Water, three times	7 to 8	45

It is thus seen that there is no regular relationship between the grain-size of the initial austenite and its susceptibility to coarsening. Further, if the grain-size of the initial austenite were the only determining factor in any one steel, then the increase in coarsening would follow similar trends for all prior heat treatments; that is to say, a treatment which produced more coarsening at 1900 degrees Fahr. (1040 degrees Cent.) would produce still more at 2000 degrees Fahr. (1095 degrees Cent), and the curves in Figs. 4, 6 and 7 would have about the same shapes. The departure of the quenched specimens from the remaining regularity shows that this is not the case. One seems warranted in concluding that the heat treatment has affected the degree of dispersion of the grain growth inhibitors in the steel, and that this effect is superimposed on the effect of initial austenite grain-size, the two effects combining to influence the coarsening temperature.

Effect of Hot Working on Coarsening Temperature

In considering these influences on the coarsening temperature, it is now an easy step to the next factor, namely hot working. From the nature of the changes which may be effected by hot working, one would be prepared to find some influence on the coarsening temperature. Hot working effects a certain amount of grain refinement,

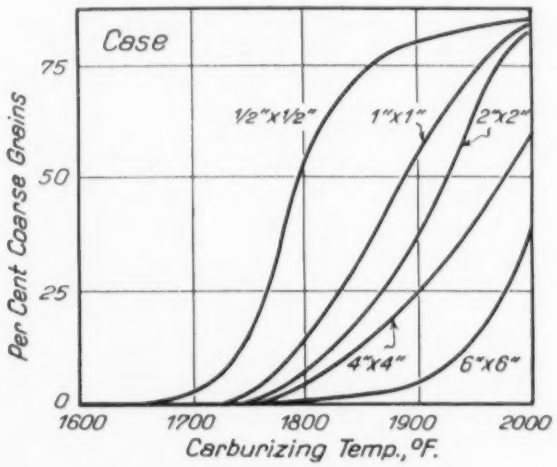


Fig. 8—Steel No. 3 (Table I) Showing the Results of Prior Hot Working on the Case.

and it is readily conceivable that it would affect the degree of dispersion of some grain growth inhibitors. At any rate, effects on the coarsening temperature have been encountered, and a series of tests is summarized in Figs. 8 and 9. Steel Number 3 of Table I was used in the form of 6x6-inch billets. Five pieces of billet were heated to 2300 degrees Fahr. (1260 degrees Cent.) and held for several hours. One piece was then removed from the furnace and merely cooled in air, another was forged to 4 x 4 inches and cooled in air, and the others to 2x2, 1x1 and ½x½ inches respectively. Samples from these five sizes were then tested for austenite grain coarsening in the way already described, and the results are plotted in Figs. 8 and 9. In the carburized case, Fig. 8, it is seen that as the hot working continues (i.e. as the cross section of the steel is reduced) there is a progressive lowering of the coarsening temperature and increased susceptibility to grain growth. It is perhaps worth pointing out that at 1700 degrees Fahr. (925 degrees Cent.) (the usual temperature for the McQuaid-Ehn test) the ½x½-inch size began to coarsen, whereas all other sections were still fine, i.e.,

they were still at the size of the initial austenite grains. In the uncarburized core, Fig. 9, hot working has likewise lowered the coarsening temperature, though it will be seen that the two largest sizes behave similarly, and that the three smallest sizes behave rather alike, there being considerable disparity between the two groups.

Austenite Grain Growth and the McQuaid-Ehn Test

It becomes possible now to assemble the known laws of austenite grain growth, and to point out their bearing on interpretations of the

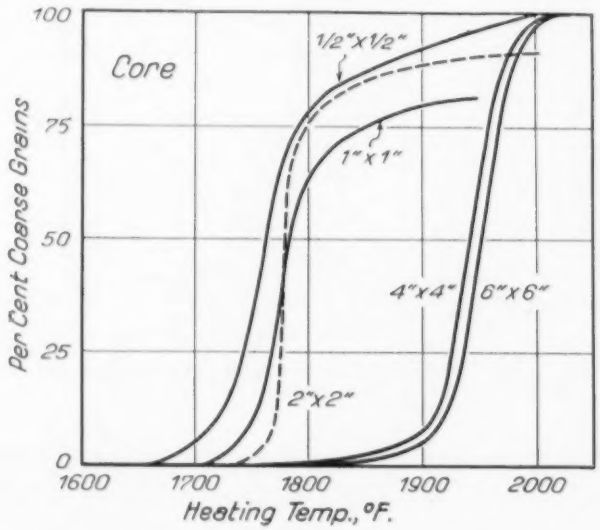


Fig. 9—Steel No. 3 (Table I) Showing the Results of Prior Hot Working on the Core.

McQuaid-Ehn test. (These laws apply more particularly to the lower carbon steels, perhaps under 0.50 per cent carbon.)

1. Austenite may coarsen anywhere in the range of temperature from about 1500 to about 2100 degrees Fahr. (815 to 1150 degrees Cent.). If the McQuaid-Ehn test shows grain-size in the range from number 5 to number 8, the steel is probably still at its initial austenite grain size at that temperature (1700 degrees Fahr.). If the test shows coarser grains (number 1 to number 4), the chances are that the steel has coarsened.

2. After carburizing in pack carburizer, the austenite grain-size of the case may be finer than the austenite grain-size of the core (14). This means almost invariably that the case is at its initial austenite grain-size, while the core has coarsened.

3. The austenite grain-size of the case may be coarser than the austenite grain-size in the core (15). Such a circumstance means that the case has coarsened while the core is still at its initial austenite grain-size.

4. Normalizing may make a finer McQuaid-Ehn grain-size, particularly in alloy steels such as nickel, nickel-molybdenum and nickel-chromium steels. In this case the normalizing has caused the ferrite grains in the steel before carburizing to be finer, and as a result (15) the initial austenite grains are finer. The coarsening temperature has not been reached.

5. Cold work may lead to a finer McQuaid-Ehn grain size by the same mechanism as in Item 4, i.e., by making finer ferrite grains.

6. Normalizing may on the other hand also lead to coarser McQuaid-Ehn grain-size. Normalizing causes a lowering of the coarsening temperature, and if in any particular steel the normalizing has so lowered the coarsening temperature that the steel coarsens at 1700 degrees Fahr. (925 degrees Cent.) while the unnormalized steel fails to coarsen, then the normalized steel will be coarser in the McQuaid-Ehn test than the unnormalized. The same comments apply to any heat treatment that lowers the coarsening temperature.

7. Hot work may make a somewhat finer McQuaid-Ehn grain-size, depending partly on the finishing temperature. The mechanism is the same as in Item 4, i.e., it makes finer ferrite grain-size, while the coarsening temperature has not been reached.

8. Hot work may lead to coarser McQuaid-Ehn grain-size. Hot work frequently lowers the coarsening temperature, and if it lowers it to below 1700 degrees Fahr. (925 degrees Cent.) while the original steel fails to coarsen at this temperature, then the hot-worked steel will be coarser in the McQuaid-Ehn test than the original steel.

ACKNOWLEDGMENT

It is a pleasure to mention the assistance and co-operation received from the author's associates at the Illinois Steel Company.

Bibliography

1. E. Heyn, "Mikroskopische Untersuchungen an tiefgeätzten Eisenschliffen", *Mitt. Kgl. Tech. Versuchsanstalt*, Vol. 16, 1896, p. 310-331. A study of the shape of etching pits in mineral crystals. Although others had

found large crystals of iron, this was the first statement that the small grains of which metallic bodies are composed were all crystalline.

2. C. A. Edwards and L. B. Pfeil, "The Production of Large Crystals by Annealing Strained Iron", *Journal*, Iron and Steel Institute, 1924—I, Vol. CIX, pp. 129-147. Describes in detail their technique for making very large crystals of iron.
3. H. Mark, M. Polanyi and E. Schmidt, "Vorgaenge bei der Dehnung von Zink Kristallen", *Zeitschrift fuer Physik*, Vol. 12, 1922, p. 58-116. A study of the ductility in different crystal directions in large single crystals.
4. C. A. Edwards and L. B. Pfeil, *Journal*, Iron and Steel Institute, 1925. A steel was treated to produce a range of grain sizes, by straining and annealing, and the pieces with their various grain sizes tested for their tensile properties. There was found to be great regularity in the variation of tensile properties with grain size.
5. R. Arrowsmith, "Effect of Grain Size on Extension at the Yield Point", *Journal*, Iron and Steel Institute, 1924-II, Vol. CX, p. 317. Iron of three markedly different grain sizes was found to display great differences in the amount of extension at the yield point in the tensile test.
6. W. H. Bassett and C. H. Davis, "A Comparison of Grain Size Measurements and Brinell Hardness of Cartridge Brass", *Transactions*, American Institute of Mining and Metallurgical Engineers, February, 1919. Various grain sizes produced in brass by rolling and subsequent annealing, and the resulting materials examined for Brinell hardness.
7. N. A. Ziegler, "Properties of Large Iron Crystals", *Transactions*, American Institute of Mining and Metallurgical Engineers, February, 1930. A study of the maximum magnetic permeability of iron, showing that the permeability increases with the grain size.
8. It is well known that 5 per cent silicon steels will usually show lower electrical watt loss when they consist of coarse grains than when they consist of fine grains.
9. A. Sauveur, "Note on the Crystalline Growth of Ferrite Below Its Thermal Critical Range," *Proceedings*, International Association Testing Materials, 1912, Vol. II. Low-carbon steel was deformed by a Brinell impression and then heated to 650 degrees Cent. (1200 degrees Fahr.). The irregular grain growth observed led to the announcement of the "critical strain hypothesis", thus beginning a new field of investigation in grain growth in metals.
10. Chappell, "Recrystallization of Deformed Iron", *Journal*, Iron and Steel Institute, 1914-I, p. 460-498. A study of grain growth after cold working and reheating. The amount of cold work was regulated by using tapered tensile test bars. The variations in grain growth were demonstrated with great clarity.
11. Mathewson and Phillips, "Recrystallization of Cold-Worked Alpha Brass on Annealing", *Transactions*, American Institute of Mining and Metallurgical Engineers, Vol. 54, 1916, p. 1-50. A demonstration that the effects of cold work and annealing of brass are rather similar to those shown in iron by Sauveur and by Chappell.
12. Carpenter and Elam, "Crystal Growth and Recrystallization in Metals", *Journal*, Institute of Metals, 1920, No. 2, Vol. XXIV, p. 83-131. Detailed study of crystal growth in aluminum and in antimony-tin alloy. Noteworthy for its illustrations of the mechanism of crystal growth, by photomicrographs after successive heatings.
13. H. W. McQuaid and E. W. Ehn, *Transactions*, American Institute of Mining and Metallurgical Engineers, Vol. 67, 1922, p. 341-391. A study

of carburized steels, showing a relation between the microstructure of a piece cooled slowly from the carburizing and the hardness of a companion piece quenched from the carburizing. This is the first recorded work on the effect of austenite grain size in hardening.

14. E. C. Bain, "Factors Affecting the Inherent Hardenability of Steel", TRANSACTIONS, American Society for Steel Treating, Vol. XX, No. 5. Nov., 1932, p. 385-428. A beautiful exposition of the hardening of steel, summarizing his own previous data and adding new data on the effects of alloys, of grain size and of "abnormality."

15. M. A. Grossmann, "On Grain Size and Grain Growth", TRANSACTIONS, American Society for Steel Treating, Vol. XXI, No. 12, Dec. 1933, p. 1079-1104. A study of the mechanism of the formation of austenite grains and the manner in which they coarsen.

DISCUSSION

BRADLEY STOUGHTON:[1] It seems to me that some comment should be made, perhaps by Dr. Grossmann or others, in a more extended way on the importance of their findings concerning the influence of the extent of reduction in rolling from the ingot on the qualities of the steel. It was formerly believed that a piece of steel must be reduced at least four to one from the ingot before it reached its highest quality. Then a committee of some of the steel companies a few years ago, under the auspices of the American Society for Testing Materials, studying the matter very carefully came to the conclusion that while the properties of ultimate strength and elastic limit were not greatly altered, the ductility was improved if the reduction was three to one instead of four to one. A good many consumers did not change their practice, however, and still require a reduction of four to one. Even as late as the past two weeks there was something published in the technical literature, as many of you have noticed, in which another investigator pointed out the conclusion he had reached that three to one reduction was better than four to one reduction. Dr. Grossmann has certainly brought out some striking evidence in favor of less reduction with consequent less liability to coarsening.

Author's Reply

I would like to say just a word about this very interesting point brought up by Professor Stoughton. It seems to me that there are several things to be kept in mind in considering the amount of reduction and one of them is the directional properties. As Charpy showed a long time ago, impact tests at least, and assuredly other properties as well, vary depending on the direction in which the test piece is cut, and as shown in Charpy's work, the greater the amount of reduction the greater was the increase in impact value when the test piece was cut with the direction of rolling, and the greater the decrease in impact value if the test piece was cut across the direction of rolling, and it may well be that for some purposes a three to one reduction would be better than

[1]Professor of Metallurgy, Lehigh University, Bethlehem, Pa.

four to one if test pieces and tests were made transversely, whereas the reverse might be true if the tests were made longitudinally.

In regard to the matter of coarsening, there is no question that in some grades of steel, at least (although we believe not in all grades), the coarsening temperature is substantially lowered by extensive hot-working.

GENERAL RELATIONS BETWEEN GRAIN-SIZE AND HARDENABILITY AND THE NORMALITY OF STEELS

By E. S. Davenport and E. C. Bain

Abstract

This paper deals with steels which, although of similar composition, yet develop different characteristic austenite grain-size at various heating temperatures alike for all. These individual grain growth tendencies correlate well with hardenability and certain mechanical properties, the coarser grained steels being, in general, deeper hardening and less tough.

Methods for estimating grain-size in various types of steel are described as well as the factors controlling the grain growth characteristics. It is believed that dispersed particles are the primary cause of restricted grain growth and that these obstructions, acting also as transformation nuclei, accelerate transformation and hence decrease hardenability; for this reason hardenability and austenitic grain-size at the hardening temperature are closely correlated.

A manifestation of the nucleation effect of dispersed particles is found in low carbon steels, case carburized and slowly cooled. An explanation of the so-called "normal" and "abnormal" structures found in the hypereutectoid case is offered.

Introduction

TO MANY experienced in hardening steel there is nothing really new in the thought that the characteristics of steel may vary from lot to lot in a manner not to be explained by the corresponding variations in the composition of the steels as reflected in the ordinary chemical analysis. Many authentic instances have been recognized in which entirely satisfactory and utterly unsuitable heats of equally sound steel have, upon analysis, been found to differ in composition only within the allowable limits. Even now, with improved and extended methods of analysis, it is questionable whether or not the

A paper presented as part of the Grain-Size Symposium of the Sixteenth Annual Convention of the Society held in New York City the week of October 1, 1934. The authors, who are members of the Society, are associated with the Research Laboratory of the U. S. Steel Corporation at Kearny, N. J. Manuscript received July 5, 1934.

precise behavior of a steel could be predicted from a knowledge of its composition alone; usually the analysis is sufficient for most purposes, but in instances wherein the hardening characteristics must be very closely controlled, factors other than the usual reported analysis must be recognized.

At one time it was regarded as necessary to melt certain grades of tool steel in crucibles even though equally clean steel of the same nominal composition could easily be produced from electric furnace or open-hearth ingots. The reason was that the electric steel when heat treated in the same standard way simply did not yield tools with the same depth of hardness, or the same toughness, found in the crucible steel. There has undoubtedly been some reluctance toward the recognition of factors, other than the composition of the steel in the customary terms of C, Mn, Si, Cr, Ni, P, S, etc., which may influence the hardening characteristics, and hence the final properties, of steel as much as does a fairly large change in composition. However, it is now possible to sum up a large part of the observations with respect to the extra-analytical variables in steel in a simple way.

Briefly, steels of substantially identical composition differ with respect to the extent to which they coarsen at the various temperatures to which they may be heated in working and in hardening; those steels which do not coarsen are in general less deeply hardening and tougher even when hardened through and through in small sections. The tendency to resist coarsening at high temperatures can be developed in carbon steels as desired by conditioning the molten steel ready for casting into ingots, so that for a specified subsequent schedule of workings and heatings the steel will possess, within limits, the desired grain-size. For a large majority of purposes the tougher steels are to be preferred, though not for all; where easy machinability is of paramount importance the structures resulting from the coarser grain may be more suitable.

Before attempting a detailed discussion of the effects of grain-size upon the properties of steel it is clearly necessary to set forth precisely the meaning of grain and grain-size insofar as this enquiry is concerned. With respect to hardenability, "timbre," "personality," grain-size control, etc., the grains under discussion are those of the *austenite,* at the heating temperature, from which the final structure under observation is formed. By grain-size is therefore meant the grain-size of the austenite established just prior to the significant cooling operation. In general this is the final heating

operation for hardening by quenching but it may refer to a prior heating, as in normalizing, if the matter of "inherited" influence is under discussion.

What is specifically not meant by grain-size is a ferrite grain-size, for example, in a low carbon steel, which influences drawing properties; or the size of the roughly triangular pearlite patches often found in a medium carbon steel. In eutectoid pearlite, it is difficult to imagine what could be considered as a grain: the single lamina of ferrite or carbide?, the portion of pearlite from a single transformation nucleus?, or the ill-defined portion which resulted from an entire austenite grain? In this discussion there will be no occasion to refer to any grain other than that of austenite; henceforth, grain-size means the grain-size of the austenite established just prior to the cooling which set up the final structure under consideration.

Estimation of Grain-Size

In correlating the properties of steel with the degree of coarsening of the austenite at the heat treating temperature, it is first of all necessary to have a method of measuring reliably the actual grain-size established in the significant heating operation. The grain structure must be well marked out in a specimen at ordinary temperature, and therefore no longer austenitic, which can be examined under the microscope, for clearly it is impracticable to examine the specimen at the heating temperature just prior to cooling when the austenite, with its grain-size upon which in part the subsequent behavior depends, is actually established. This is not entirely easy, for in some cases, the structure formed during the transformation occasioned by cooling does not in any way reflect the grain-boundaries of the parent austenite. For example, in either the martensite (or its tempered products) or in the pearlite of approximately eutectoid composition it is frequently quite impossible to discern the location of the austenite grain boundaries. In certain slowly-cooled, low carbon steels the ferrite-pearlite patchwork seen in the microscope bears no apparent relation to the pattern of the austenite grains. Nevertheless, in practically any steel the austenite grain structure can, with some expenditure of effort, be marked out in an unequivocal manner.

The most important consideration in making an estimation of austenitic grain-size is the choice of the temperature and time at temperature employed in the test, so that it duplicates closely the condi-

tions under which the significant heating will be carried out in the commercial use of the material. The necessity of this practice will become more evident as discussion proceeds.

The following methods have been found entirely satisfactory in determining the grain-size in the various carbon ranges.

Distinctly Hypereutectoid Steels. In most cases, particularly with plain carbon steels, it suffices merely to cool the specimen in air after heating substantially according to the schedule contemplated or observed in the commercial operation under consideration. The specimen, which may vary from perhaps a quarter-inch section to a small billet, may be cooled at any moderately slow rate, and about each austenite grain a continuous envelope of carbide will be formed which offers sufficient contrast to the pearlite to make grain-size estimation quite feasible.

If the steel contains considerable alloy or if it is only slightly hypereutectoid it may be necessary to transfer the specimen, after heating and prior to final cooling, to another furnace or bath at a temperature just above A_1 for several minutes to insure the formation of the intergranular carbide network. The ideal cooling rate is then that which will form a black-etching pearlite of finest lamellae; in the larger sections an oil quench is often effective. In some alloy steels martensite may form, in which case contrast between network and groundmass may be developed either by tempering and etching with alcoholic nitric acid or by etching in boiling sodium picrate solution. The austenite grain of a hypereutectoid steel is marked out in typical manner in Fig. 1.

Moderately Hypoeutectoid Steels. In many cases in which the volume of proeutectoid ferrite is inherently comparatively small the same general procedure as for hypereutectoid steels may be effectively employed. The network marking out the austenite grain will, in this case, be ferrite and the etching technique is appropriately selected. Likewise, if the steel is only slightly hypoeutectoid a brief halt in the cooling at a temperature just above A_1 will aid in forming a more nearly continuous network of contrasting ferrite. Fig. 2 shows the austenite grain marked out in typical manner in a hypoeutectoid carbon steel of about 0.45 per cent carbon. When the carbon content of the steel is much lower with respect to the eutectoid composition good results may usually be secured by halting the cooling at a temperature just below A_3 and then cooling rather more rapidly than for

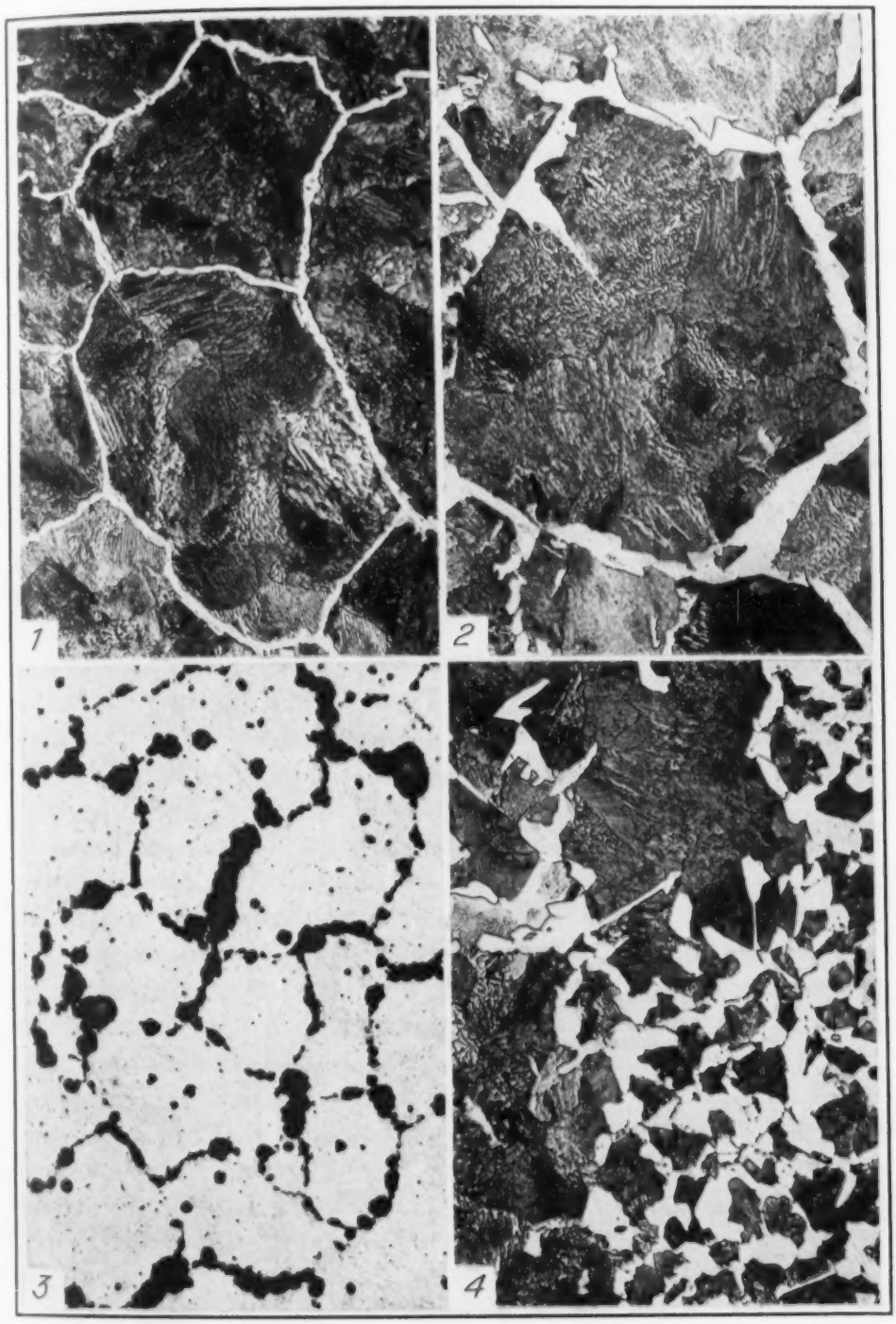


Fig. 1—Microstructure of Hypereutectoid Steel Showing Austenite Grain-Size Marked Out by Carbide Network. × 300.

Fig. 2—Hypoeutectoid Steel Showing Austenite Grain-Size Marked Out by Ferrite Network. × 300.

Fig. 3—Eutectoid Steel, Cooled at Rate Just Less than Critical Quenching Speed, Showing Austenite Grain-Size Marked Out by Fine, Nodular Pearlite Network. × 125.

Fig. 4—Mixed Grain-Size in Hypoeutectoid Steel as Marked Out by Ferrite Network. × 300.

eutectoid compositions, e.g., by an oil quench; the network is thus rendered more narrow and clean cut.

Low-carbon Steels. It is rare indeed that any interest is manifest in the austenite grain-size of ordinary low carbon steels other than steels to be case-carburized in use; such steels should, of course, be carburized in the test for grain-size and treated thereafter as hypereutectoid steels. When, however, it is desired to observe this feature it may be accomplished in the following manner. A moderately small sample is preferable; it is cooled from the appropriate heating temperature to somewhat below the A_3 temperature where it is held for a few minutes to establish a ferrite network. It is then quenched as rapidly as possible. If not too low in carbon a subsequent tempering will produce a dark etching groundmass, particularly if the boiling sodium picrate solution is used as an etching agent. In some cases, when the carbon content is very low, only the texture of the etched surface serves to mark out the network and considerable patience and experience may be required to estimate the grain-size.

Eutectoid Steels. With no proeutectoid constituent to form in the grain-boundaries of the austenite, the grain is often obliterated with the transformation, either to pearlite or to martensite, or its tempered products. However, one characteristic of the transformation to fine pearlite serves to mark out the grain boundaries with remarkable clarity; the transformation invariably begins in the grain boundaries at a number of points acting as nuclei. Accordingly, if the steel is cooled at a rate just less than the critical quenching speed, the microscopic specimen shows clearly where the boundaries of the parent austenite lay. For any eutectoid carbon steel a specimen of suitable size may be chosen for quenching so that at some point in its section the cooling must have been at the optimum rate. The etched section then reveals the grain-size as in the photomicrograph of Fig. 3; the groundmass is martensite and the nodules in the grain boundaries are fine pearlite. In rare cases of rather high alloy steels this type of demarcation of the grains may have to be secured by cooling the specimens quickly into a lead bath for a determined interval at a temperature some 100 to 200 degrees Fahr. (50 to 100 degrees Cent.) below the A_1 to permit the desired degree of transformation to fine pearlite to occur, after which the specimen is quenched into water to insure that the remainder of the specimen becomes martensitic.

Another procedure for grain demarcation in eutectoid steels is described by Grossmann[1] as having been devised by Miss M. Baeyertz. In this case a special polishing and etching method applied to the wholly martensitic specimen causes a satisfactory degree of contrast between the groundmass and the material in the vicinity of the former grain boundaries of the austenite.

All schemes for marking out the grain boundary pattern of austenite ready to be suitably cooled depend upon causing a specific reaction to occur at the boundaries different from that in the groundmass of the grains; the differential attack of the etching then develops the required contrast for the examination of these patterns.

When the representative specimen of the steel has been so prepared that the grain boundaries of the parent austenite are set forth with adequate contrast, the next step is the estimation of the grain-size. This is most directly accomplished by counting the grains in a definite known area; with experience, the mere observation of the image of the etched surface at a determined magnification on the ground-glass permits an approximate estimation of grain-size. A somewhat more reliable method is the comparison of the image with a series of standard photomicrographs of known grain count. Such standards for comparison are available, either actual photomicrographs or idealized hexagonal patterns. The numbered charts of the American Society for Testing Materials are based upon the number of grains per square inch of the image at a magnification of $\times$ 100. These grain-size designations are now in wide use for steel although in nonferrous work the number of grains per square millimeter, or the average area of each grain on the section, is more often used as a basis of comparison. The following table shows the range of grains per square inch at $\times$ 100 covered by each number and the mean value.

Grain-Size Index, N	Number of Grains Per Square Inch at $\times$ 100, n		
	Mean	Max.	Min.
1	1	1.5	..
2	2	3	1.5
3	4	6	3
4	8	12	6
5	16	24	12
6	32	48	24
7	64	96	48
8	128	..	96

Obviously the basis for this system is, $n = 2^{N-1}$

[1]M. A. Grossmann, "On Grain-Size and Grain-Growth." TRANSACTIONS, American Society for Steel Treating, Vol. XXI, No. 12, 1933, p. 1079.

Accordingly, if larger or smaller grains than those encompassed in this series are encountered they can easily be designated in a consistent manner; as for example:

Grain-Size Index, N	Number of Grains Per Square Inch at × 100, n		
	Mean	Max.	Min.
−2	⅛ or 0.125	0.17	0.08
−1	¼ or 0.25	0.35	0.17
0	½ or 0.5	0.7	0.35
9	256	384	192
10	512	768	384
11	1024	1536	768

It may, at times, be convenient to utilize higher magnifications than × 100 for fine-grained structures and lower magnifications for the coarsest structures. In that case the comparison may still be made directly with the standard charts except that for each doubling of the magnification above × 100 the apparent grain-size must be raised by two numbers. Thus if a structure at × 400 matches the No. 2 chart then its grain-size is 6. If the match is with No. 3 at a magnification of × 50 then its grain-size is really 1. Similarly a shift in grain-size of one number corresponds to a change in magnification by the factor $\sqrt{2}$ or 1.4.

Grain-size estimation involves certain matters of judgment due to the random cutting of the component grain envelopes. Inevitably even large grains will be sectioned occasionally through a corner in such a manner that the traces of its boundaries will enclose a very small area. Accordingly, no note should be made of an occasional small grain. On the other hand, it is quite usual to find steels in which a range of grain-sizes actually exists, or, more often, two dominant distinct sizes of grain. If more than two or three contiguous small grains are found together then there must necessarily be really finer grains present and this fact should be noted. As well described by Grossmann, the formation of large grains occurs by the gradual absorption of the small ones first formed as the ferrite and carbide disappear, and accordingly there are frequent occasions when a mixed grain must inevitably be encountered. The present authors have found it convenient to note these mixed structures in one of two ways. When there appears to be a full representation of sizes over a definite range the designation is 3 to 4, 4 to 6, etc., and if the coarser grains predominate the smaller number may be underscored; when two dominant sizes appear as in the usual intermediate stage of coarsening it is useful to indicate the proportion, for example, 60%

7, 40% 3 or 30% 6, 70% 2. Thus in the hypoeutectoid steel of Fig. 4 the examination of many fields, such as the one in the photomicrograph at × 300, indicated a grain-size rating of 25% 1, 75% 6.

A common criterion of "fineness" and "coarseness" has traditionally been employed by steel workers, which consists of a comparison of the textures of fractures. It appears that the fracture, if it results from breaking in the brittle manner, i.e., without permanent deformation, is made up of transgranular facets which in general originate from the austenite grains. Certainly the texture of a martensitic fracture correlates with austenitic grain-size, and probably for a restricted range of carbon content the "brittle" fracture reflects austenitic grain-size even in conditions other than martensite.

Grain-Size and Grain Growth Tendency

A definition of the word "grain-size" for present purposes has already been set up. It describes a single, definite state of affairs as found in a specimen. Different steels at the same temperature after the same time interval may possess different grain-size, or the same steel after equally long heatings at different temperatures will develop grains of different size. It has been observed that some steels coarsen at the customary, relatively low heating temperatures, while others may remain fine-grained when likewise heated. A broad classification may thus be made, if finer discrimination is unnecessary, by designating the steels as having either marked grain growth tendency or restricted grain growth tendency, but such categories are not useful unless the temperature range is understood to be restricted in a recognized manner. Unfortunately, the terms are abbreviated sometimes to merely "coarse-grained" and "fine-grained" with no mention of temperature. However, where steels are to be heated by a well-recognized schedule no ambiguity results from using the terms "inherently fine-grained" or "inherently coarse-grained" steel.

The complete description of the grain-growth tendencies of a steel, even if it is to be conventionally heated for a reasonable period, is only fully set forth by a table, or chart, of the grain-size acquired in this heating at a series of temperatures. A series of such figures for a number of steels selected at random appears in Table I.

The evidence is quite complete as to the inherently fine-grained steel being shallow hardening as compared with the inherently coarse-grained steel when identically treated. As an example, the compari-

Table I

Grain-Size—Temperature Characteristics of Several Steels

Temp. Degrees Fahr.	Temp. Degrees Cent.	Steel A	B	C	D	E	F	G	H	I	J	K
2100	1150	.....	.....	.....	.....	.....	.....	{80% 2 {20% 3	1—	.	.	.
2000	1095	.....	.....	.....	.....	.....	.....	2 to 3	{50% 2 {50% 3	.	.	.
1950	1065	1 to 4	2 to 6	.....	.....	.....	.....	.....	.....	.	.	.
1900	1040	.....	.....	.....	.....	.....	.....	{50% 3 {50% 4	3 to 4	.	.	.
1850	1010	2 to 5	3 to 6	.....	.....	.....	.....	.....	{90% 4 {10% 5	2	3	4
1800	980	.....	.....	1 to 3	3 to 4	3	4 to 5	3 to 4	6 to 7	.	.	.
1750	955	2 to 6	2 to 6	1 to 3	3 to 4	3	4 to 5	.....	.....	.	.	.
1700	925	.....	.....	2 to 3	3 to 5	3	4 to 5	4	{80% 8 {20% 6	3	4	5
1650	900	3 to 7	4 to 8	2 to 3	{90% 3 {10% 7	3 to 4	4 to 5	.....	.....	.	.	.
1600	870	.....	.....	3	6	{90% 3 {10% 7	4 to 5	5	8	.	.	.
1550	845	4 to 8	8	3	6 to 7	6	5	.....	.....	4	6	6
1500	815	.....	.....	5 to 6	8	7	5 to 6	6	8+	.	.	.
1450	790	.....	.....	.....	.....	.....	5 to 7	.....		.	.	.
1400	760	.....	.....	.....	.....	.....	7 to 8	6	8+	6	7	7

son curves of Fig. 5 based upon substantially identical compositions illustrate this very well.

If now, grain-size is a really significant factor in the hardenability, and not merely an attendant circumstance, then a single steel should be capable of exhibiting an increasing depth of hardening after identical quenches if the austenite grain-size is caused to be-

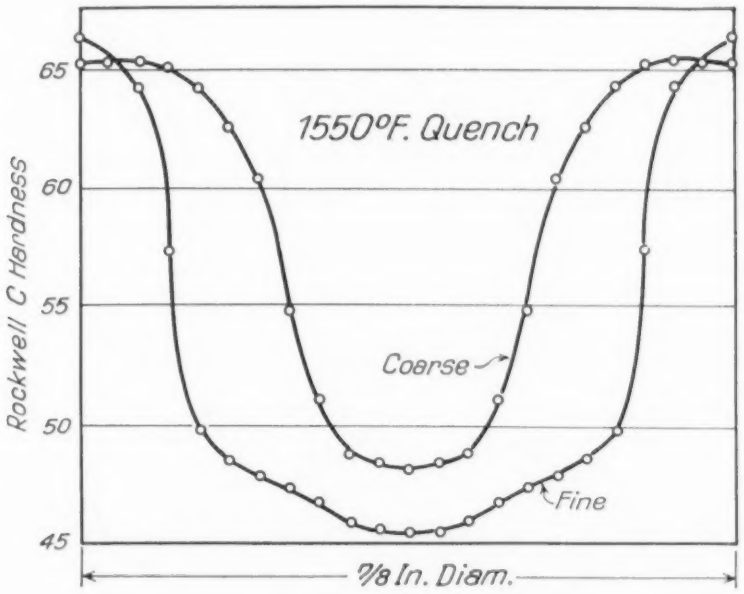


Fig. 5—Curves Showing Depth of Hardening in Identically Quenched ⅞-Inch Rounds of Fine-Grained and Coarse-Grained Steels of Substantially Identical Composition.

come successively coarser. Accordingly, before considering the steels of inherently different characteristics it may be well to examine critically the effect of grain-coarsening upon individual steels.

Correlation of Actual Grain-Size and Hardenability of a Steel

It is an extremely old observation that a steel as quenched from higher and higher temperatures into water or brine became successively more and more deep hardening. In most quarters the influence of the higher heating temperature to bring about a more rapid quench, and so to force the critical quenching speed to deeper zones, was emphasized whereas possible changes in the steel itself developed by the higher heating temperatures were often minimized

or neglected. It appears now that this emphasis was in error. The point may be investigated rather easily.

If a number of 4 or 5-inch lengths, for example, of 1-inch rounds be cut from the same bar they may be assumed to be essentially identical. In order to establish a still closer degree of uniformity they may be given a high temperature normalizing treatment or better still, an oil quench. Suppose then that all the speci-

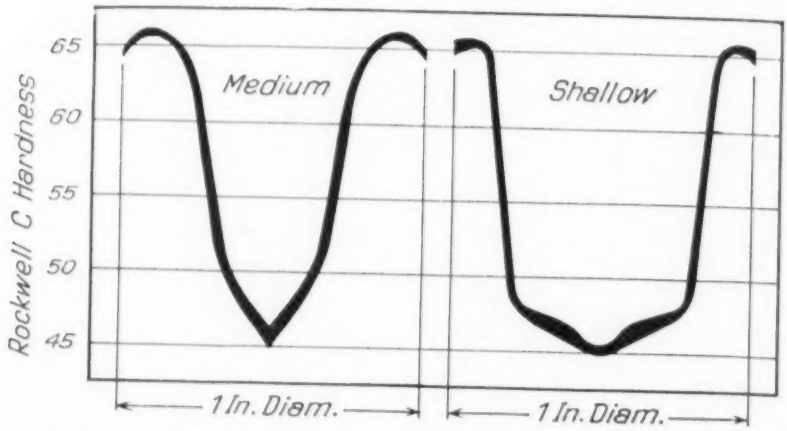


Fig. 6—Effect of Quenching Temperature Alone Upon the Hardenability of Two Steels Having Each a Constant Grain-Size. Hardness Values for the Various Quenching Temperatures (1600-1375 Degrees Fahr.) all Fall Within the Narrow Black Bands.

mens are heated to some 1600 degrees Fahr. (870 degrees Cent.) for precisely the same time; they will then have the same austenite grain-size. They may now, however, be quenched from different temperatures into brine by transferring each to a lead bath successively adjusted to each of a series of temperatures—say between 1375 and 1600 degrees Fahr. (745 and 870 degrees Cent.), for a few minutes before quenching.

The results of such an investigation are shown in the hardness plots of Fig. 6 in which both a shallow hardening and a medium hardening steel were heated to 1600 degrees Fahr. to establish the grain-size and then quenched from 1600, 1525, 1450 and 1375 degrees Fahr. (870, 830, 790, 745 degrees Cent.). The single hardness plots overlap and the envelope of all is shown by the solid black band. The width of this band is little greater than the inherent error of the method and there is no definite trend in the individual plots encompassed in the band. Hence one may conclude that the actual quenching temperature, so far as water or brine is concerned, plays a minor role indeed in hardenability, but that the heating temperature which

establishes the grain-size is the significant factor. The work of Scott[2] on rates of cooling in quenching would indicate such a state of affairs, and these findings were therefore predicted.

By a change in procedure, it is possible to quench from the same temperature, a series of specimens differently coarsened but as we

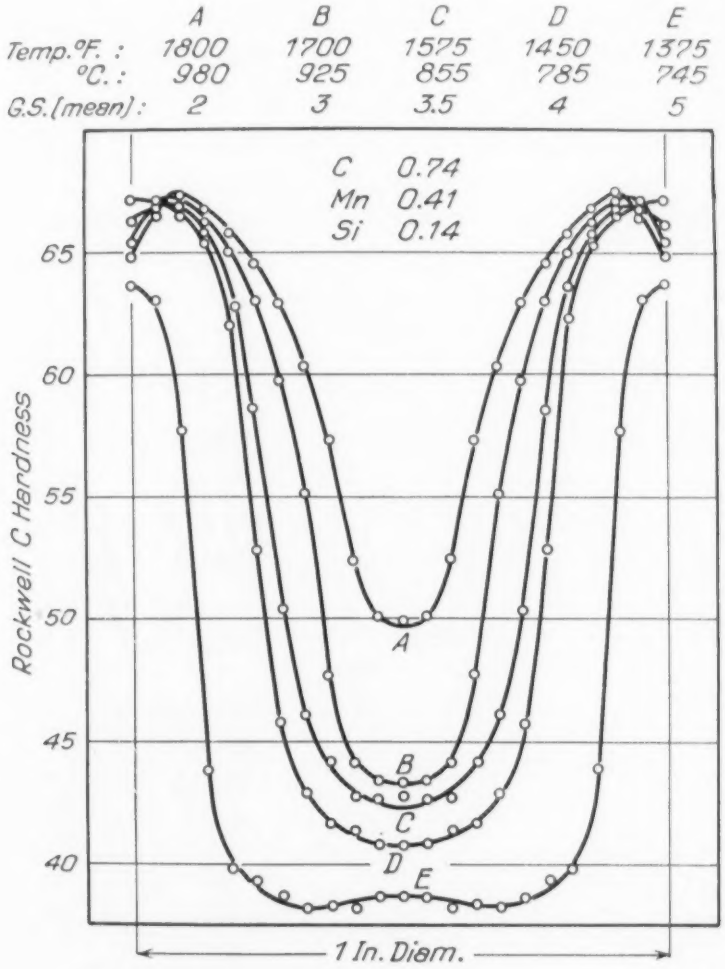


Fig. 7—Hardenability of a Single Steel as Affected by Increasing Austenitic Grain-Size.

have seen this is not entirely necessary since actual quenching temperature has so little effect in a brine quench.

A typical example of the effect of a gradually coarsened grain upon hardenability is shown in Fig. 7. The actual grain-size is tabulated for the corresponding hardness plot. A very simple demon-

[2]Howard Scott, "Some Problems of Quenching Steel Cylinders." TRANSACTIONS, American Society for Steel Treating, Vol. XXII, No. 1, 1934, p. 68.

stration of increased hardenability resulting from a coarsened grain is shown in Fig. 8 in which both the fracture and the polished and etched sections of the specimen as well as the hardness plots are included.

The hardenability of a steel is a function of its rate of transformation, in the early stages, at the temperature at which this rate is a maximum, usually in the vicinity of 1000 degrees Fahr. (540 degrees Cent.). This relation has been discussed in considerable detail elsewhere[3] and will not be further treated here. It is, however, significant that the rate of transformation of a steel at constant subcritical temperature, some degrees below the A_1 temperature, depends upon the grain-size established in the austenite just prior to the rapid cooling to the temperature of actual transformation. In Fig. 9 is shown the actual course of transformation in a series of specimens of a 1.17 per cent carbon, 0.30 per cent manganese steel with austenite grain-size as indicated. It is interesting to note that a change in grain-size from No. 8 to No. 2 to 6 retards the transformation as much as would result from the addition of perhaps one per cent manganese at constant grain-size.

From observations of this sort, which have been made on a wide variety of steels, it appears that there can be little question of the validity of the correlation between established grain-size, rate of reaction, and the attendant hardenability of a steel.

Inherent Grain-Growth Tendency

As far as one may judge, the high carbon steels of older production, particularly those made with the greatest care and marketed as tool steel, were often prone to coarsen if heated very much above the minimum temperature which sufficed for hardening. It is not to be wondered at that the heat treater with a good eye for temperature, and thoroughly experienced with respect to the time limitations for heating, was envied and proud of his skill, for it is probable that temperatures much above about 1450 degrees Fahr. (790 degrees Cent.) were often dangerous and at the same time, full hardening could scarcely have resulted from a heating below about 1400 degrees Fahr. (760 degrees Cent.). The range was narrow,

[3] E. C. Bain, (Campbell Memorial Lecture), "Factors Affecting the Inherent Hardenability of Steel," Transactions, American Society for Steel Treating, Vol. XX, No. 11, 1932, p. 385.

E. C. Bain, "On the Rates of Reaction in Solid Steel." *Transactions*, American Institute of Mining and Metallurgical Engineers, Vol. 100, Iron and Steel Division, p. 13.

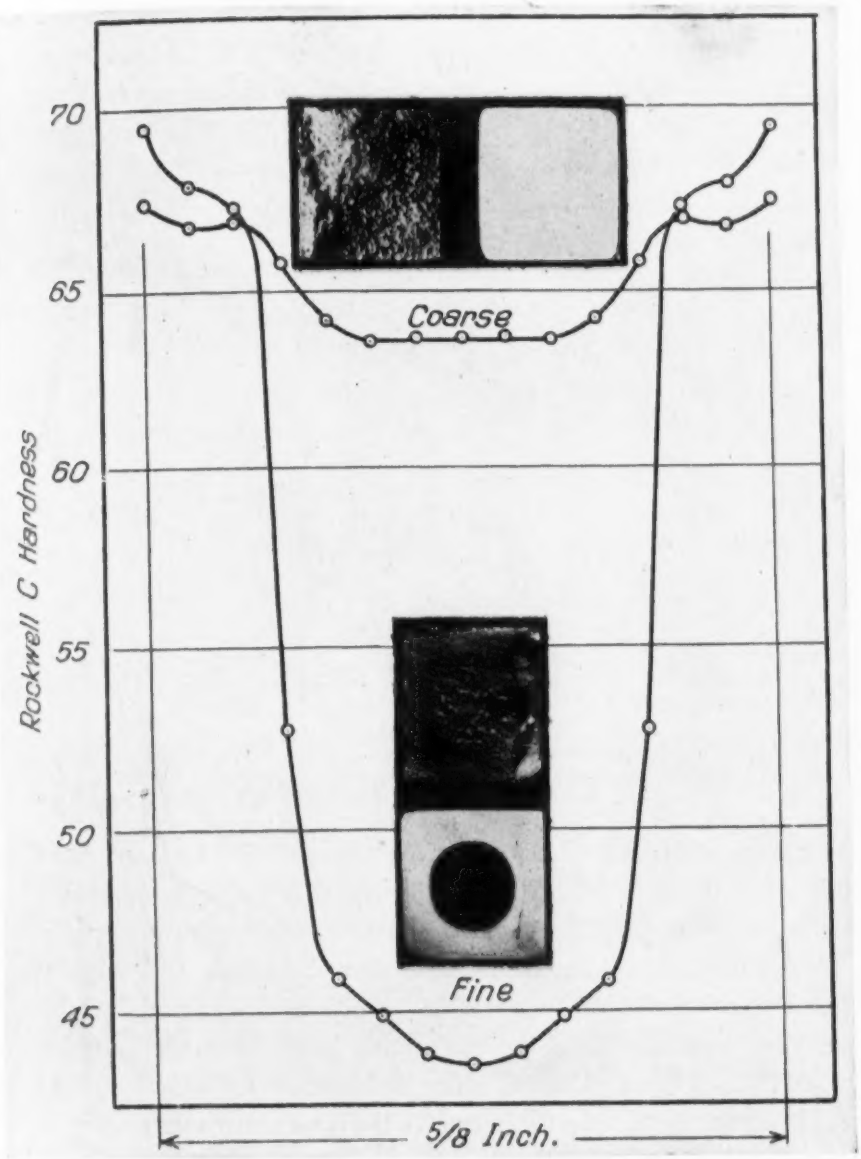


Fig. 8—Hardenability of a Single Steel as Affected by Coarsening Treatment. G. S. of 5 in the Deep-Hardened Specimen Established at 1800 Degrees Fahr. (980 Degrees Cent.); G. S. of 8 in Shallow-Hardened Specimen Established at 1550 Degrees Fahr. (840 Degrees Cent.); Both Quenched from 1375 Degrees Fahr. (745 Degrees Cent.).

and, if the upper limit was exceeded, the resulting tool would be excessively brittle, if indeed it did not crack in the water quench. A marked improvement resulted from the use of higher manganese content which permitted the use of an oil quench; this steel even though somewhat coarsened, was less disposed to crack in the milder

quench. The temperature range for hardening these inexpensive oil-quenching steels carrying higher manganese was still somewhat narrow, for the toughness of the coarsened steel was not invariably adequate. Other alloying elements were employed such as tungsten, chromium, and vanadium, which restricted the coarsening and widened the safe heating range; the toughness was definitely improved,

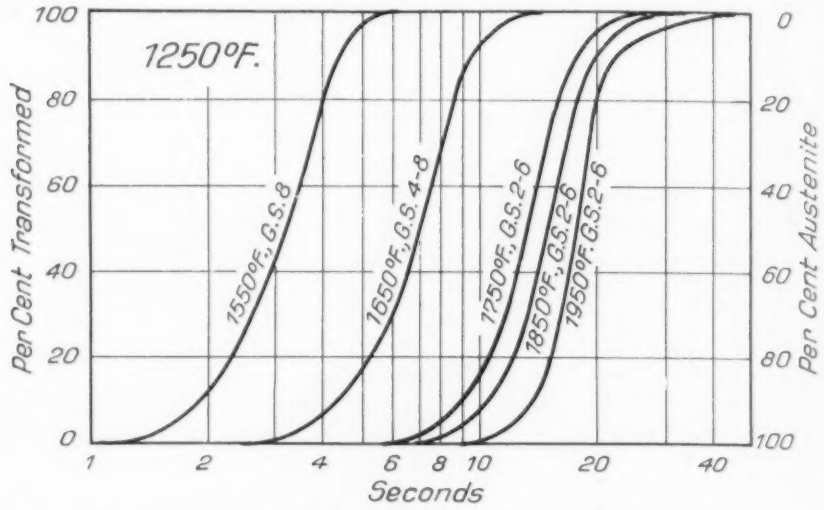


Fig. 9—Transformation Rates at 1250 Degrees Fahr. (675 Degrees Cent.) in a Single Steel as Affected by Change in Austenitic Grain-Size.

particularly with vanadium additions to either the water-hardening carbon steel or the oil-hardening intermediate manganese steel.

Particularly with respect to the water-quenching carbon steels the observation was long ago made that the steels to which a small proportion of vanadium was added were often more shallow-hardening, and quite properly the extra toughness was attributed to this circumstance. Thus, although the subject appears not to have engaged much critical attention up to the last few years, the parallel between hardenability and the grain-coarsening tendencies did not entirely escape notice even many years ago.

Shallow hardening is readily explained without recourse to a consideration of the grain-size in all cases in which the retention of fine grain structure in the austenite accompanies the presence of fairly large amounts of strong carbide-forming elements, as in tungsten- and molybdenum-bearing steels, heated to a temperature too low to dissolve nearly all of the carbide present. The resulting low carbon content of the austenite in itself explains the low hardenability although the presence of the carbide particles appears also to

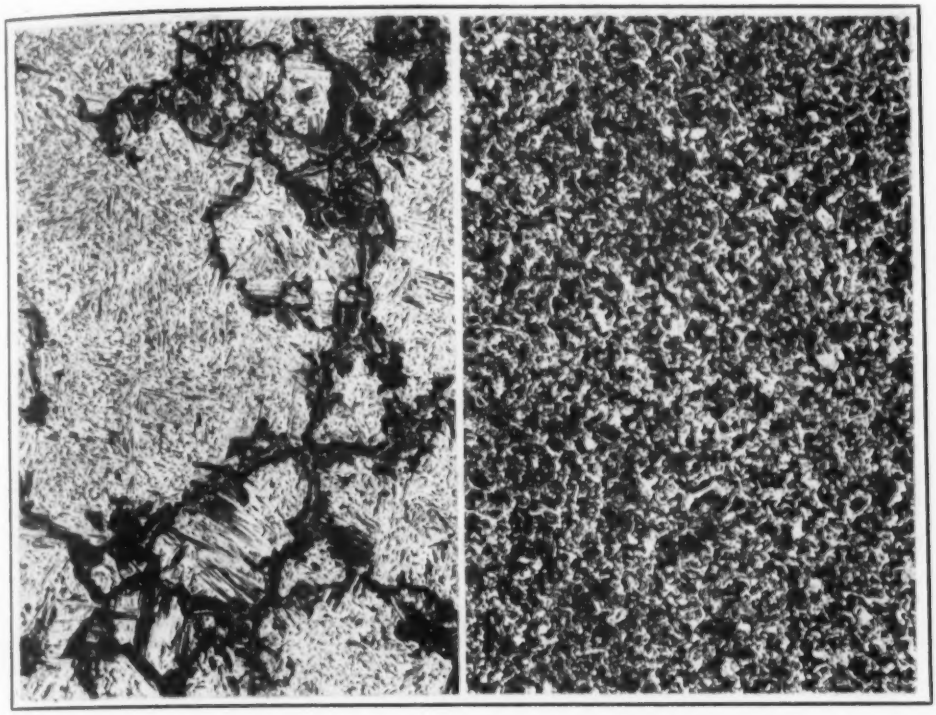


Fig. 10—Grain-Size Characteristics in Unkilled and Aluminum-Killed 0.39 Per Cent Carbon Steels as Identically Quenched from 1600 Degrees Fahr. (870 Degrees Cent.). Left—Coarse-Grained, Unkilled; Right—Fine-Grained, Aluminum-Killed. Microstructure in the Interior of Quenched 1-Inch Rounds. × 100.

restrict grain growth. It is more interesting, however, to consider the disparity in hardenability between two very similar plain carbon steels which under the same circumstances develop different sizes of austenite grain.

A very simple example of this comparison is afforded in the steels developing, as identically quenched, the contrasting structures of Fig. 10. These two steels originated from the same experimental heat of a 0.40 per cent carbon steel. Half of the steel was cast without any deoxidation. Mild effervescence occurred and since proper rimming action naturally could not be established, most of the ingot was merely a sponge. However, a portion of the ingot at the bottom was fairly sound and this was forged to 1-inch rounds. The remainder of the heat was killed with aluminum; the proportion added being about 8 ounces per ton. The analysis of the bars is as follows:

	C	Mn	P	S	Si	O_2[4]	N_2
Killed	0.39	0.09	0.007	0.012	0.002	0.010	0.0026
Un-killed	0.37	0.11	0.006	0.010	0.004	0.015	0.0043

[4] "Oxygen" was determined by the high temperature hydrogen-reduction method of Larsen and Brower. Quite clearly no correlation between grain-size and this readily dissolvable oxygen is to be expected.

Presumably about two points of carbon were lost as CO in the effervescing. In the unkilled steel there were indications of some iron oxide inclusions but in the killed steel none of them were found. As heated to 1600 degrees Fahr. (870 degrees Cent.) the killed steel developed an austenite grain-size of No. 7 while the unkilled steel had largely coarsened to a No. 1 size, and as quenched in the form of disks ¼ inch thick, the unkilled steel hardened through and

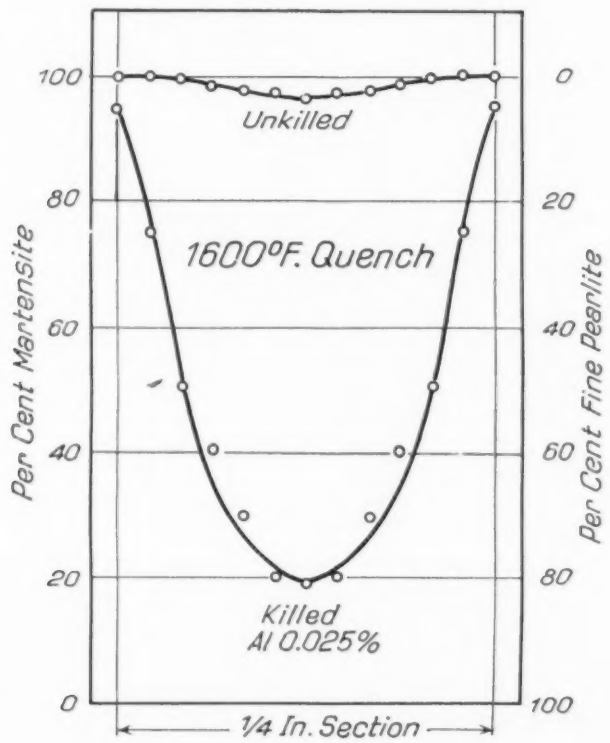


Fig. 11—Hardenability of the Unkilled and Aluminum-Killed 0.39 Per Cent Carbon Steels of Fig. 10 as measured by the Proportion of Martensite and Fine Pearlite in ¼-Inch Disks Quenched from 1600 Degrees Fahr. (870 Degrees Cent.).

through with only 2 per cent fine pearlite in the center while the killed steel hardened only near the surface as indicated in Fig. 11. The profound change in hardenability resulting from the addition of 0.025 per cent aluminum implies a distinct change in actual transformation rate. Confirming this, the actual course of the transformation of the two steels at 1275 degrees Fahr. (690 degrees Cent.) as heated alike to 1650 degrees Fahr. (900 degrees Cent.) is shown in Fig. 12. The transformation rate of the killed steel is roughly twice that of the unkilled steel.

Similarly in a tool steel composition range two steels of similar composition exhibit a markedly different grain-growth pattern and a corresponding disparity in depth of hardening. The analytical data on the two steels are as follows:

	C	Mn	Si	P	S	Ni	Cr	Al
Coarse-Deep	1.03	0.23	0.16	0.014	0.017	0.10	0.06	0.001
Fine-Shallow	1.08	0.23	0.08	0.012	0.006	0.03	0.12	0.015

The grain established at various temperatures is shown in the following tabulation:

Temp. Degrees Fahr.	Temp. Degrees Cent.	Coarse-Deep	Fine-Shallow
1500	815	5 to 6	7
1550	845	3	6
1600	870	3	90% 3 10% 7
1650	900	2 to 3	3 to 4
1700	925	2 to 3	3
1750	955	1 to 3	3
1800	980	1 to 3	3

The comparative depth of hardening resulting from two temperatures of heating, 1500 and 1600 degrees Fahr. (815 and 870 degrees Cent.) in brine-quenched 1-inch rounds is shown in Fig. 13. A difference in physical properties in the two steels, as hardened throughout, will be mentioned subsequently.

The decrease in hardenability resulting from the restriction of grain growth may be offset in a steel of the same carbon content by the introduction of some element, such as manganese, which, by its retarding influence upon the transformation, contributes strongly to greater hardenability. As an example of this manganese effect, the behavior of three nominally 0.72 per cent carbon steels,[5] the one carrying higher manganese, is set forth in the hardness plots of Fig. 14. The grain-size established in the three steels at the three heating temperatures is indicated in the following tabulation:

Temp. Degrees Fahr.	Temp. Degrees Cent.	Coarse-Deep	Fine-Shallow	Fine-Deep (Mn)
1475	800	3 to 4	6 to 7	6 to 7
1550	845	2 to 3	6	6
1650	900	2 to 3	50% 2 50% 7	90% 6 10% 3

[5]These steels were regular commercial heats received from the Carnegie Steel Company and are typical of the controlled grain-size product.

The higher manganese steel, at once fine-grained and deep hardening, possesses the advantages in physical properties of both tendencies and much of the merit of alloy steel clearly depends upon this

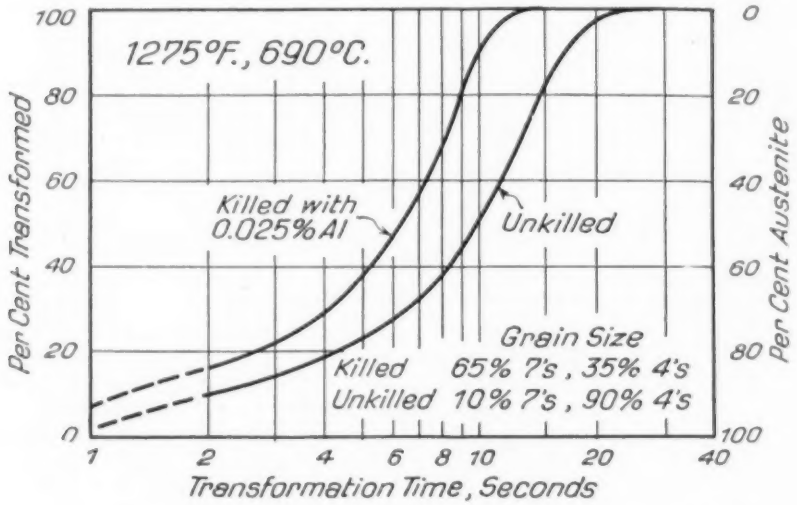


Fig. 12—Transformation Rates at 1275 Degrees Fahr. (690 Degrees Cent.) of the Unkilled and Aluminum-Killed 0.39 Per Cent Carbon Steels of Figs. 10 and 11.

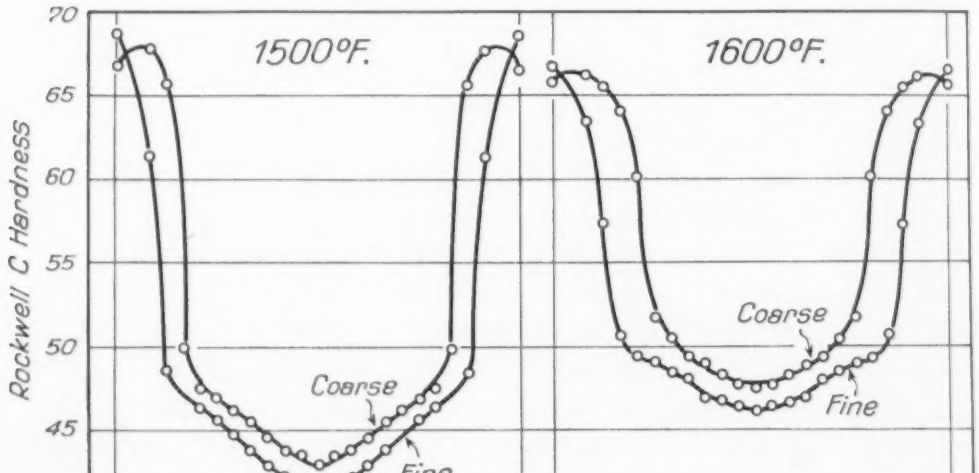


Fig. 13—Hardenability of Coarse-Grained Deep-Hardening and Fine-Grained Shallow-Hardening Tool Steels As Quenched from 1500 Degrees Fahr. (815 Degrees Cent.) and 1600 Degrees Fahr. (870 Degrees Cent.) in the Form of 1-Inch Rounds.

combination of characteristics. In the examples above, the increase in the manganese content and a slightly higher carbon content almost precisely compensated for the tendency toward shallower hardening

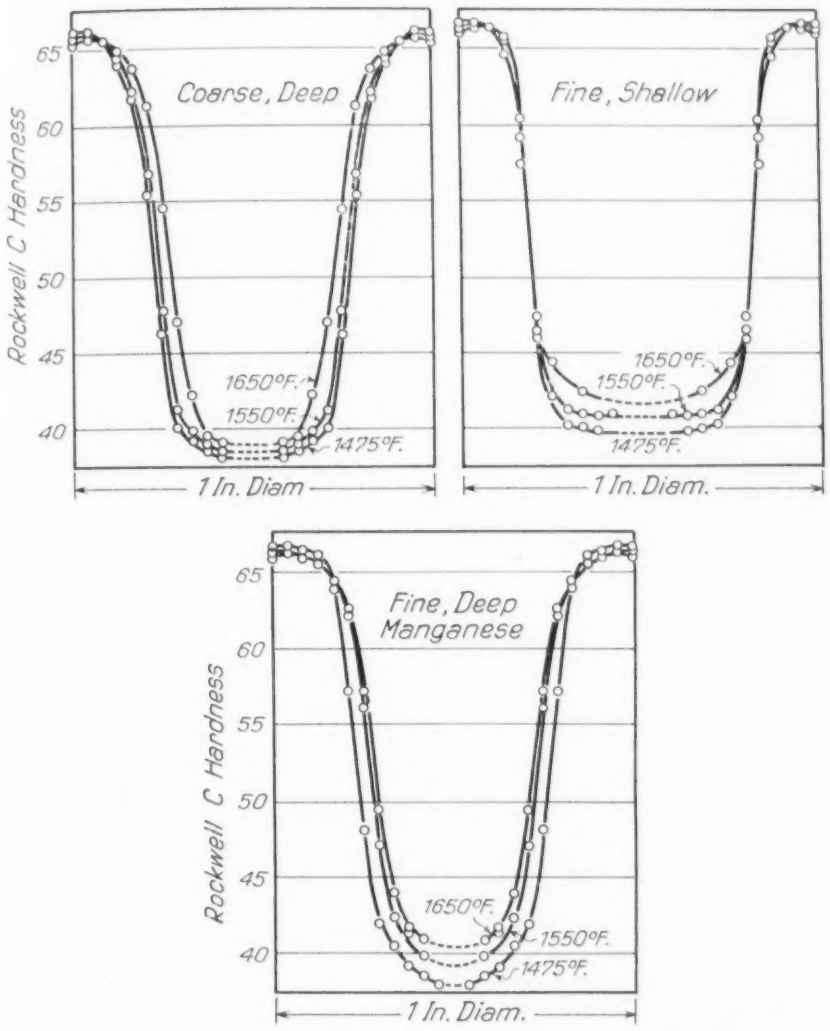


Fig. 14—Hardenability of Three 0.72 Per Cent Carbon Steels As Quenched from 1475 Degrees Fahr. (800 Degrees Cent.), 1550 Degrees Fahr. (845 Degrees Cent.) and 1650 Degrees Fahr. (900 Degrees Cent.) in the Form of 1-Inch Rounds.

through restriction of grain growth. A larger increment of alloy element would have been necessary had a greater disparity in grain-size at quenching temperature existed, in order that the hardenability of the fine-grained manganese steel should have been equal to that of the coarse steel.

General Influence of Austenitic Grain-Size Upon Mechanical Properties of Steel

Attention has already been called to the traditional practice of

hardening steel from the very minimum heating temperature which, in the time interval set by various conditions, would just suffice for the solution of sufficient carbon to develop the necessary hardenability. This practice is, of course, commendable and entirely sound when toughness is of primary importance. The examination of a number of successfully treated tools which have functioned well indicates that a fairly large proportion of the carbide remained undis-

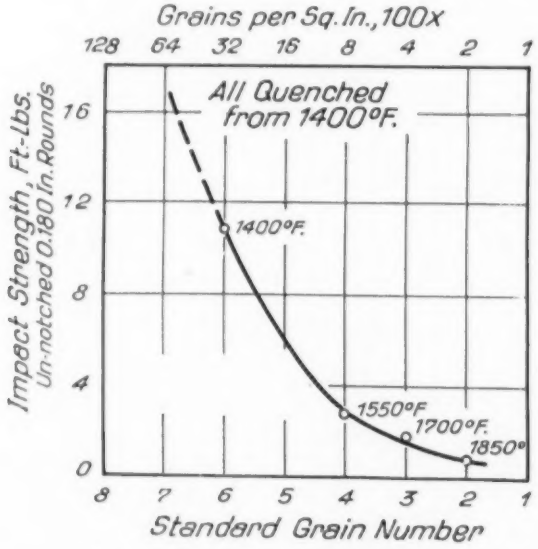

Grains per Sq. In., 100x
128 64 32 16 8 4 2 1
All Quenched from 1400°F.
1400°F.
1550°F
1700°F.
1850°
Impact Strength, Ft.-Lbs. Un-notched 0.180 In. Rounds
16 12 8 4 0
8 7 6 5 4 3 2 1
Standard Grain Number

Fig. 15—Influence of Austenitic Grain-Size Upon Impact Strength of a 0.85 Per Cent Carbon Steel. Grain-Size Established at the Indicated Temperatures but All Specimens Quenched from 1400 Degrees Fahr. (760 Degrees Cent.) and Tempered to 50 Rockwell C Hardness.

solved and that, therefore, these carbide particles are imbedded in a tempered (formerly martensitic) matrix rather low in carbon. Beyond question this condition is very satisfactory in many applications; but in many other applications it has been found preferable to secure a high proportion of the carbon in solution in the martensite and then, by tempering carefully, to re-form a fine dispersion of carbide in precisely the proper degree of fineness. Obviously, this demands higher heating temperature for quenching or longer time-interval at temperature, both of which procedures encourage grain-growth. Indeed, it appears that such grain-growth as occurs in the higher carbon, heat treating steels begins just when the major portion of the carbide disappears by going into solution. If, therefore, a large part of the carbon content of the steel is to participate in the solution-reprecipitation reaction which is responsible for the strength-

ening by heat treatment, it is essential to restrain undue grain growth in some manner. When this is accomplished better combinations of plasticity with strength or hardness result.

The direct influence of mean grain-size alone upon the impact strength of a steel is shown in Fig. 15. In this case un-notched specimens were employed so that the results would have some significance in relation to actual uses of steel, in which good design carefully avoids the extraordinary stress gradients which exist in notched bars. The specimens were originally fully martensitic throughout and then tempered to exactly 50 Rockwell C for the entire lot of specimens over the range of grain-sizes. All the 0.180-inch rounds were quenched from 1400 degrees Fahr. (760 degrees Cent.) although the coarsened grain was secured by preheating at higher temperatures as indicated.

The product of the transformation of coarse grains is clearly of different character from that of finer grains. It is thought that the stresses set up during transformation are greater in the case of coarse austenite because the absolute dimensions of the Widmanstatten plates are greater and the origin of the stress would appear to be in the volume differences between the original austenite and the transformed fragments. If the internal stresses—compression and tension—within the steel were greater in the coarse and smaller in the fine specimens, then it is most probable that the impact strength would likewise differ and that the lower stressed material would offer greater resistance to rupture under the blow of the hammer. This subject has been more fully discussed elsewhere[6] and it suffices for the immediate purpose to note that tempered martensite is the tougher the smaller the grain-size of the austenite from which it formed.

Having established this trend for the grain-size effect in individual steels in which the austenite grain is regulated by heating temperature, the next point of interest would be the effect of grain-size as controlled in the several steels themselves, as heat treated identically. This relationship is well illustrated in three 1.0 per cent carbon steels whose grain-size and impact values for the wholly martensitic condition are shown in the following tabulation.

---

[6]E. S. Davenport, E. L. Roff, and E. C. Bain, "Microscopic Cracks in Hardened Steel, Their Effects and Elimination." TRANSACTIONS, American Society for Metals, Vol. XXII, No. 4, 1934, p. 289.

Heating Temperature	Coarse Steel G. S. No.	Impact	Fine Steel G. S. No.	Impact	Very Fine Steel G. S. No.	Impact
1500 degrees Fahr. (815 degrees Cent.)	5 to 6	4.6	7	8.6	8	11.2
1550 degrees Fahr. (845 degrees Cent.)	3	3.6	6	6.2	6 to 7	10.0
1600 degrees Fahr. (870 degrees Cent.)	3	3.5	{90% 3 {10% 7	3.8	6	6.0

This is quite representative of the correlation between grain-size tendency and impact strength of steels of the same composition. In the heating time employed the carbon was by no means wholly dissolved at 1500 degrees Fahr. (815 degrees Cent.); some remained after the 1550 degree Fahr. (845 degree Cent.) heating but not sufficient to prevent marked grain growth.

The Special Characteristics of Very Coarse Austenite Grains in Inherently Fine-Grained Steel

The direct correlation between grain-size, hardenability and toughness of the fully hardened steel holds over such a wide range of conditions and is so useful, that it is perhaps unimportant to take up certain very unusual conditions in which the grain-size does not appear to act exactly in accord with the simple concept thus far set forth here. On the other hand, the seemingly anomalous behavior of the extreme grain-size range may cast some light upon the action of grain growth inhibitors upon transformation characteristics; for this reason a reference is believed desirable at this point.

The general grain growth pattern of the easily coarsened steels is quite simple and not at variance with that of many other simple solid solutions which have been studied. The temperature of most marked grain growth (in the customary short time interval usually employed in heat treating) is generally in the vicinity of 1525 to 1575 degrees Fahr. (830 to 855 degrees Cent.) at which temperature the marked coarsening generally results in a grain-size in the vicinity of No. 3, or more broadly 2 to 5. Heating temperatures above this range, even up to perhaps 1850 degrees Fahr. (1010 degrees Cent.), do not succeed in establishing much coarser grains; perhaps the same grain or even a finer one may result at the higher temperature if the heating is very rapid. Usually the coarsest grain established will not exceed a No. 2. The main point is that as soon as the carbide is largely dissolved the marked coarsening is immediately accomplished,

and these grains are of quite uniform size and not fine; accordingly they have, as would be expected, a minimum tendency to grow.

Other steels so produced as to carry an effective grain growth inhibitor, behave in a quite different manner. The coarsening temperature is very high indeed, but the resulting grain-size, when this high heating temperature is actually applied, is vastly coarser than

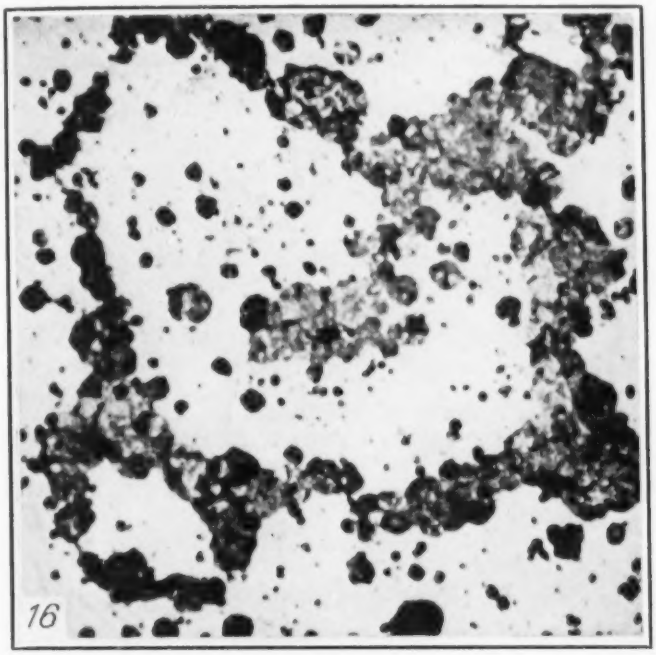


Fig. 16—Photomicrograph of Coarsened Shallow-Hardening Steel Quenched at a Rate Just Less Than the Critical Quenching Speed. Note Fine Pearlite Nodules at Many Transformation Centers Within Grains in Addition to Those at Austenite Grain-Boundaries. × 100.

in the steels with less effective restrainers of grain growth. The actual grain-size is sometimes so coarse as to correspond with a No. 5, each grain being perhaps $\frac{1}{10}$ inch in diameter after coarsening is finally brought about at, say, 2100 degrees Fahr. (1150 degrees Cent.).

If now, such a coarsened specimen is quenched, it might be expected to be very deep hardening, and very brittle. Actually it is neither. When the fine grain structure is extremely persistent nearly to these unusually high heating temperatures, the steel remains *effectively* fine-grained in spite of its actual coarse austenite grain. The depth of hardening is no greater than would be expected from a grain-size of perhaps 4 or 5, not deeper, indeed, than that of a

similar composition in a deep-hardening steel as quenched from 1500 degrees Fahr. (815 degrees Cent.).

It has previously been shown[2] that such unusually coarsened steels begin to transform, not alone in the grain boundaries, but at many nuclei throughout the grains, having, indeed, as many transformation nuclei as a much finer grained austenite which begins its transformation solely at the grain boundaries. The existence of many transformation centers is well shown in the photomicrograph of Fig. 16, taken at a point within a specimen which cooled at a rate just lower than the critical quenching speed. Presumably also the transformation centers for martensite formation are likewise numerous, for the nature of the martensite is the same as that from really fine-grained austenite. On this basis it is not precisely austenite grain-size but rather its effective grain-size, or the number of transformation centers in the austenite, which actually correlates thoroughly with hardenability and toughness.

A number of observations point to this conclusion and it seems necessary to revise the usual concept of pure grain-size influence to something which is applicable to all cases. This revision is, however, a very slight one. It appears reasonable that the same particle dispersion which can increase the number of austenite grains can also, when very effective, exert a large influence on the number of centers from which the transformation upon cooling can begin to occur. In steels, then, with only mild grain growth inhibition the transformation begins only at the grain boundaries, whereas, in the steels in which the crystallization centers are very effective in restricting austenite grain growth, the transformation is not restricted initially to grain-boundaries but may begin at many points, even within the body of the individual grains.

It appears then that the relatively ineffective particles are those which dissolve, or which coalesce, at relatively low temperature, and which therefore are unable to carry the marked grain growth range up to high temperature whereat really gigantic grains are formed. If this be correct, clearly such weak grain growth inhibitors could also not serve as nuclei for the transformation upon cooling. It may well be that hardenability best correlates with the temperature of marked grain growth rather than with grain-size unless the grain-size be restricted to that which developes at relatively low heating temperature, as discussed in the early portion of this paper.

To sum up, one may conclude that the cause of high initial rate

of reaction of a steel of given composition is a function of the capacity of the austenite to afford a relatively large number of transformation centers or nuclei. In the general case, these centers of transformation occur in the austenite grain-boundaries and accordingly a fine-grained austenite must always accompany a relatively shallow-hardening effect, assuming, of course, similar composition. The nuclei for transformation, as they become more and more numerous, appear also to force the marked grain growth in the austenite to higher and higher temperatures. Accordingly, over a very wide range of conditions hardenability parallels grain-size for similar compositions. When, however, the particles which inhibit grain growth are so effective that only at very high temperature will the grains grow, then the growth becomes exaggerated and in this case the particles serve also to act as transformation centers, irrespective of the boundaries of the enormously large grains. Accordingly, hardenability in similar steel compositions may quite properly be spoken of as a function of grain-size only over a definite, though wide, range of grain-sizes; beyond these limits the hardenability becomes less and less strictly correlated with actual grain-size and should be thought of as a function, rather, of *effective grain-size* or nucleation capacity. Since in no practical application is the exaggerated grain growth referred to allowed, the revision of the concept does not alter the conclusions already drawn by many investigators from grain-size data.

The Mechanism of Grain-Growth

The sequence of events in a steel heated through the critical temperature range and the mechanism of the establishment of the austenite grain structure has been lucidly discussed by Grossmann[1] with particular emphasis upon the behavior of low carbon steel during carburization. The observations and conclusions of the present authors are so essentially like those of Grossmann that it seems unnecessary to repeat the account here. Since, however, the present discussion applies principally to the higher carbon steels, the present views of austenitic grain establishment are set forth very briefly:

1. Upon heating, the first points of austenite formation are all at the interface between carbide and ferrite; the earliest austenite nuclei to form are those which are located in regions where the manganese, or other similar element, is fortuitously at higher concentration, where the carbide, if lamellar, is in the thinnest plates or if

spheroidal, is in the finest particles. These first pools of austenite must be of very small size and exceedingly numerous. Thereafter any large regions of ferrite or carbide are absorbed as the austenite "colonies" grow. Every opportunity then arises for the more favored orientations to grow and to absorb smaller austenitic grains less favorably located for growth, for austenite is formed above its own recrystallization temperature.

2. The temperature for the establishment of a moderately coarse, uniform grain system in pure carbon steel austenite is evidently not much above the critical temperature, for even at 1500 to 1550 degrees Fahr. (815 to 845 degrees Cent.) a structure is often formed which in itself has such a narrow range of grain-size that it remains almost unchanged if heated to even 1800 to 1850 degrees Fahr. (980 to 1010 degrees Cent.). Such steels are easily coarsened, but any exaggerated grain growth beyond this moderately coarse structure is strongly resisted.

3. Some steels set up a fine grain structure just after transformation to austenite. It is now believed that the presence of undissolved particles is the principal cause of the failure of the austenite to coarsen at these low temperatures; either non-metallic particles or special carbide particles may act in this manner. A considerable amount of some dissolved element such as molybdenum, tungsten or cobalt may serve to raise the recrystallization temperature of the austenite sufficiently to prevent coarsening. Such steels are spoken of as being reluctant to coarsen, since at ordinary heating temperatures they remain fine-grained.

4. However, if the steel acquires a moderately fine initial grain-size it then has stronger latent powers of coarsening at some higher temperature than an easily coarsened steel, for a slight gradient in temperature, strain, or composition may induce further grain growth in a fine-grained material. The laws of inhibited grain growth and exaggerated grain growth have been well worked out for tungsten metal in which particles of thoria are distributed. In most commercial steels, the finer the initial grain established, the higher the temperature at which marked grain growth will finally occur but the coarser will be the grains resulting from the exaggerated grain growth action. Thus steels in which the grain growth inhibitor is extremely effective may develop an extraordinary grain-size at 1900 to 2000 degrees Fahr. (1040 to 1095 degrees Cent.).

Thus far, the factor of time interval at heating temperature has

been intentionally omitted from the discussion on the basis that it would be understood to conform to the ranges in actual hardening practice. Actually, of course, grain growth is a time-consuming process and it proceeds more and more rapidly the higher the temperature within the range at which the grain growth may occur. There is apparently a range, for any given fundamental grain growth tendency, below which no growth occurs in any length of time, how-

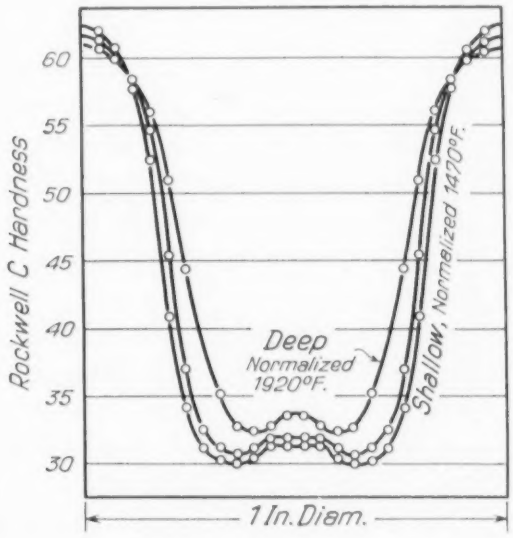


Fig. 17—Hardenability of a 0.45 Per Cent Carbon Steel as Influenced by Prior Normalizing Treatments. Shallow—Normalized at 1470 Degrees Fahr. (800 Degrees Cent.); Medium—Normalized at 1740 Degrees Fahr. (950 Degrees Cent.); Deep—Normalized at 1920 Degrees Fahr. (1050 Degrees Cent.). All Quenched for Hardening from 1490 Degrees Fahr. (810 Degrees Cent.).

ever long. Obviously, as stressed by Grossmann, when the interest centers in the grain-size acquired at a temperature within the lower part of the grain growth range of a steel, it becomes important to note that the grain established in several hours may be utterly unlike that achieved in, for example, the 20 or 30 minutes heating time for proper hardening. There will be occasion later to refer to this consideration again.

While it is thought that the paragraphs above introduce most of the major factors involved in grain growth, there are unquestionably several minor factors which on occasion act to present a more complicated picture of grain-size and hardenability. When the initial distribution of carbide particles is such as to retard their

solution in the austenite, the grain-size resulting from short heating periods at low temperatures may be unexpectedly fine and the hardenability extraordinarily low, for the austenite is, in effect, of low carbon content and hence shallow-hardening. To secure uniform deep hardening some hardening shops regularly quench the carbon steel work from a fairly high temperature into oil prior to hardening in order to secure uniformity and rapid carbon solution. Grossmann has shown the influence of prior treatments upon the grain-size. In Fig. 17 is shown a correlation between hardenability and the prior normalizing treatment. The lowest hardenability shown resulted from "normalizing" at 1470 degrees Fahr. (800 degrees Cent.) (G. S. No. 7); the intermediate hardenability was secured from the same quench after a 1740 degrees Fahr. (950 degrees Cent.) normalizing treatment (G. S. 25% 1, 75% 6); the deepest hardening in the same quench resulted in the steel normalized at 1920 degrees Fahr. (1050 degrees Cent.) (G. S. 1 to 3).

Usually, a steel once coarsened may be restored to the condition of more shallow hardening, when its deeper-hardening characteristics were "inherited" from a prior treatment, by merely renormalizing at a lower temperature. Occasionally a steel appears to be permanently changed with respect to its grain-growth characteristics; in such cases it is believed that the grain-growth inhibiting influence is lost through coalescence of the particles.

Unusual coincidences may arise in which the various minor factors influencing grain growth and hardenability may obscure the major factors; these rare cases should not be regarded, in themselves, as invalidating any broad principles if correctly deduced.

Controlled Grain-Size in Steel

It often happens that the first broad achievement of a technical improvement, such as an advance in metallurgical technology, precedes the recognition of its factors or underlying fundamentals in a scientific way. Such, undoubtedly, has been the case in the manufacture of steel with controlled grain-size. However, certain aspects of this practice are now becoming apparent and where recognized are being reflected in improved uniformity of product.

It has long been recognized that certain alloy steels were always fine-grained under ordinary conditions of heat treatment. This subject has been touched upon elsewhere and the brief remarks here

relate to steels which carry very little of those elements which, in themselves, contribute to maintenance of fine grain.

It becomes increasingly clear that the steels difficult to coarsen contain, at the heat treating temperature, some sort of dispersed particles which act to restrain grain growth, and what is most important, to increase the fundamental reaction rate of the austenite transformation. The principles underlying the relation of grain-size to reaction rate have already been set forth. Such particles, if they are effective, must neither dissolve nor coalesce at the heat treating temperatures, even though they should become ineffective at certain higher temperatures. The range of composition of such particles is therefore limited and it is not surprising that they are now thought to be substances which are highly refractory. McQuaid and Ehn[7] in 1922 reported that deoxidation with aluminum and carbon-free ferrotitanium resulted in steels which, as carburized, were fine-grained and of low hardenability. Deoxidation with aluminum certainly points to the formation of alumina or possibly aluminum silicate in the steel and the idea of refractory particles was therefore established very early. Further evidence serves to confirm the idea of alumina being particularly efficacious in restraining grain growth even when it is established as a fine dispersion *in situ* in solid steel, indeed, no evidence appears to oppose this view. The manufacture, however, of a fine-grained steel of high quality involves many other considerations beyond the mere addition of aluminum to the undeoxidized melt.

In brief, it is, in all probability, only the aluminum and oxygen which are still dissolved in the cast metal which are effective in accomplishing grain control, and the alumina formed at once as such is either eliminated from the metal or, at the worst, is retained in the form of particles too large to be effective. It is convenient to speak of the dispersion of what must be colloidal-sized particles, as though they were actually alumina; obviously their actual composition must, with present means of study, remain unknown, but it is not thought that any seriously misleading inferences will result from such a designation, and indeed, some of the evidence rather favors the conjecture that they are actually alumina.

---

[7]H. W. McQuaid and E. W. Ehn, "Effect of Quality of Steel on Case-Carburizing Results." *Transactions*, American Institute of Mining and Metallurgical Engineers, Vol. LXVII, 1922, p. 341.

E. W. Ehn, "Influence of Dissolved Oxides on Carburizing and Hardening Qualities of Steel." *Journal*, Iron and Steel Institute, Vol. CV, 1922, p. 157.

With a given total oxygen content in the liquid steel to which now aluminum is to be added, a wide range of end-results may be secured merely by a variation in the amount of aluminum added per unit of steel. The proportion of elemental aluminum remaining in the steel at, for example, heat treating temperature, steadily increases with increase in the total aluminum added. On the other hand, the proportion of alumina rejected between the solidification temperature and heat treating temperature (that is, the effective alumina), first increases and then decreases with greater and greater aluminum additions. From the same molten steel, then, one may secure solid steels, ready for heat treating, which contain equal amounts of effective alumina with different proportions of residual aluminum. This behavior is merely mentioned as an example of the complexities which arise in the consideraton of deoxidation by aluminum. The dissolvable oxygen in the instance given above varies in the reverse manner to the aluminum, although such oxygen appears to exert relatively very little influence upon the steel. When now the unavoidable variation of total oxygen in the undeoxidized bath is also taken into account, it is readily seen that great skill is demanded in the production of uniform steel of controlled grain-size.

Another element, while it is not so amazingly effective pound for pound, in restricting grain growth, as aluminum, is nevertheless noteworthy. Vanadium serves to maintain a fine grain over the normal heat treating range but permits a moderate but not excessive grain growth at heating temperatures somewhat higher. This feature is of definite value in certain cases. It appears that the effectiveness of vanadium has to do more with its high activity in forming a vanadium-rich carbide of low solubility than in creating a dispersion of refractory oxide particles. Clearly, the carbide behavior is best considered from the standpoint of the distribution of the carbon of the steel between iron and vanadium. In an ordinary eutectoid steel, as little as 0.20 per cent vanadium suffices to build these special carbide particles which are apparently not wholly dissolved even at 1700 degrees Fahr. (925 degrees Cent.). Such particles, since they may form very quickly upon quenching, probably act as strong transformation centers.

Normality and Abnormality

One of the early instances[7] of the recognition of factors, other

than nominal composition, influencing hardenability had to do with carburizing steels. Certain steels into which had diffused the usual amount of carbon failed to harden uniformly when treated according to the standard quenching practice; even recarburizing failed to induce the normal response and prevent soft spots. The microstructure of these carburized steels after slow cooling was sufficiently different from that of the steels which hardened well to permit a classi-

Fig. 18—Photomicrograph of "Normal" Steel. × 300.
Fig. 19—Photomicrograph of "Abnormal" Steel. × 300.

fication even though, at the time, it provided little knowledge of the underlying causes of the dissimilarity in hardenability. To the structure found in the slowly cooled specimens of properly hardenable steel the name "normal" was applied; to the structure in the steels which when quenched failed to harden normally the name "abnormal" was applied. This was, of course, a reasonable choice of names so far as these low carbon, low silicon, carburizing steels were concerned. The steel was so low in all elements that, in the sections involved, it had to possess almost maximum hardenability, including coarse austenite grain-size, in order to harden without soft spots. The two extreme types of structure to which these names were ap-

plied are now familiar to all, and are typified by the photomicrographs of Figs. 18 and 19. In general it was observed that the abnormal steels were fine-grained while the normal steels were coarser. The grain-size was not so much emphasized as the structural characteristics which clearly differ markedly in the extremes. Presumably the microscopic examination of these carburized steels was made in an effort to gather knowledge concerning their fundamental nature (responsible for the disparity in hardenability) rather than as a test, for clearly it is easier to quench the carburized piece and explore its hardness than to slowly cool, polish, etch and examine microscopically. At any rate, the grain-size in a hypereutectoid composition is easy to estimate and a great many examinations of carburized steels of many sorts have been made. In brief, one may characterize the extremes of normality and abnormality as follows:

Normal Structure: Thin, smooth carbide envelopes entirely contiguous with the finer lamellar pearlite.

Abnormal Structure: Thick carbide envelopes, with coarse lamellar pearlite, if any, separated from the network, which may be discontinuous, by a wide band of ferrite sometimes containing large spheroids of carbide.

It is interesting to speculate upon the trend of events which would have occurred if it had been customary to employ higher manganese, higher silicon alloys for carburizing. Such steels when fine-grained at the carburizing temperature would have had adequate hardenability and would have been tough as to core. If an easily coarsened steel of this composition had found its way to the consumer it would probably have been slightly deficient in toughness, even though the grain-size did not exceed that in the ordinary normal carburizing steel. Doubtless, the fine-grained steel would have been called "normal" and the coarse-grained type "abnormal." For this reason the words are badly chosen except they be restricted to the grade of carburizing steel to which they were originally applied. The use of such a word as "normal" immediately forces the question "for what purpose?" Such a question is reasonable and deserves its answer.

Before enquiring into the inherent differences between normal and abnormal steels it is well to bear in mind that carburization is a slow process requiring several hours even at 1700 to 1750 degrees Fahr. (925 to 955 degrees Cent.) while grain growth may occur very quickly if the grain growth temperature is below that employed

for carburizing. It may be well to re-read the conclusions of Grossmann on this subject:

> "In a carburizing test (McQuaid-Ehn test) for austenite grain-size, the steel initially (during the heating-up in the carburizing pot) undergoes its transformation to austenite and in so doing comes to a certain (small) austenite grain-size, usually in the range from No. 5 to No. 8.
>
> The temperature of the piece, of course, continues to rise thereafter, up to the carburizing temperature. Every steel has a coarsening temperature which, if exceeded, causes coarse grains (No. 1 to No. 2) to form. If the carburizing temperature is above the coarsening temperature, the final carburized piece will consist partly or wholly of coarse grains. But if the carburizing temperature is at any temperature below the coarsening temperature (and the latter may be high, even above 2000 or 2100 degrees Fahr.) then the austenite grain-size will remain unchanged, precisely what it was when the steel first transformed to austenite upon heating.
>
> In some steels this initial austenite grain-size is related to prior ferrite grain-size and is in these cases affected by prior heat treatment."

Let us consider, for the moment, two specimens of equal dimensions representing a normal and an abnormal steel of the same alloy content in a carburizing box ready for slow cooling, sufficient carbon having diffused into the specimens to yield an adequately thick case. Upon cooling, the proeutectoid carbide begins to form very soon near the surface and the envelope about the grains grows in thickness until the beginning of the three-phase temperature zone is reached. In nearly pure iron-carbon steels this zone is so narrow as to constitute essentially the eutectoid horizontal. Up to this point no essential difference exists in these specimens. If the two specimens were suddenly quenched the fine-grained steel would, at some higher magnification, resemble closely the coarser grained specimen at some lower magnification. At this point the divergence appears. Shortly before the normal specimen begins its simultaneous rejection of ferrite and carbide (pearlite) from various nuclei in the intergranular network, the austenite of the abnormal specimen begins its decomposition. Ferrite begins to be formed inside the cells of network and the carbon, thereby forced out of the shrinking austenite pools, diffuses to the network there to attach itself in the form of more carbide, thickening the network. Spheroids of carbide are from time to time formed in this widening inner rim of ferrite but many of these subsequently disappear and the carbon involved migrates to the

network. This process goes on slowly, so slowly indeed, that the temperature of the specimen may decrease somewhat, whereupon the remainder of the austenite forms a coarse lamellar pearlite. In the meantime the normal steel, which had undercooled somewhat more, continues to form solid masses of regular pearlite. This is a more rapid process and the temperature usually rises somewhat.

These observations serve to construct a fairly complete picture of the fundamental requisites[8] for the two types of steel:

1. An abnormal steel during transformation obviously possesses high carbon diffusivity.

2. This observed diffusivity increases rapidly with the increase in the actual temperature of the metal in the early stages of transformation. This implies low concentration of any elements such as nickel or manganese which lower A_{e1} or such as chromium or silicon which cause undercooling.

3. High temperature of initial transformation (minimum undercooling) is favored by the presence of many transformation centers (nuclei) as supplied by more grain boundary area or dispersed particles.

Similar reasoning may be applied to the normal type of transformation.

1. A normal steel possesses low carbon diffusivity relatively to the velocity of the austenite transformation in the early stages.

2. The observed low diffusivity decreases rapidly with decrease in the actual temperature of the transformation. This implies either the presence of elements which depress A_{e1} or which retard transformation and increase undercooling.

3. Depressed temperature for the beginning of transformation (undercooling) is greatly assisted by the absence of transformation centers (nuclei) and by large grains (i.e., effectively large) which offer diminished grain boundary surface.

Even a moderate difference in normality may be detected in steels of similar composition by comparison of cooling curves, which however, must be taken upon specimens of precisely equal dimensions, from which, furthermore, the heat must be abstracted at precisely the same rate. Such a comparison is shown in Fig. 20. In this case the steels, having nearly eutectoid composition, were not

[8] These circumstances were recognized by F. Duftschmid and E. Houdremont. "Ursache des anormalen Verhaltens von Stählen bei der Einsatzhärtung." *Stahl und Eisen*, Vol. 53, Dec. 1931, p. 1613.

carburized. They differ, however, in grain-size and hardenability and structural normality by virtue of the difference in deoxidation practice. The undercooling of the normal steel need really be no greater than that of the abnormal steel (although it usually is) pro-

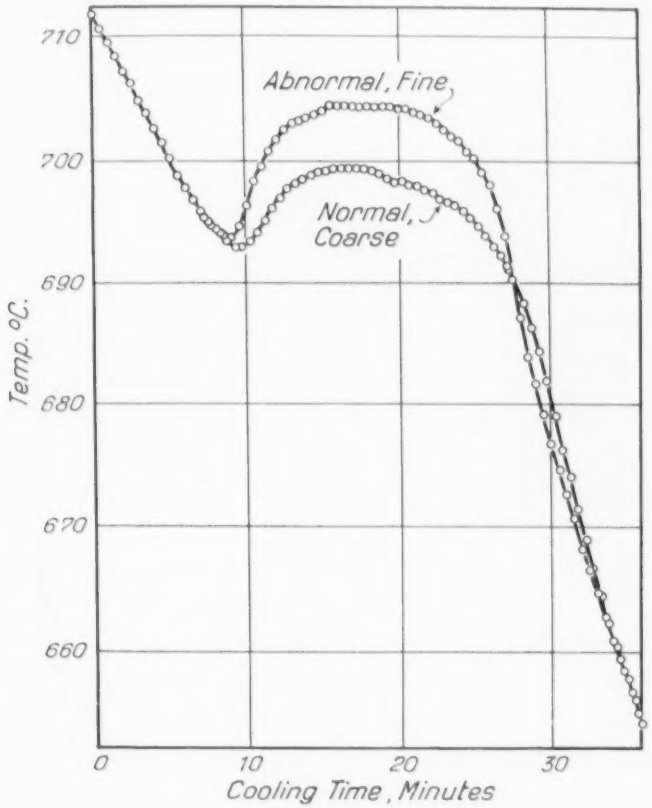


Fig. 20—Simultaneous Cooling Curves for Fine-Grained, Abnormal and Coarse-Grained, Normal Steels of Similar Composition.

vided only that the early stages of the eutectoid transformation are more rapid so that austenite decomposition is rapid relatively to the diffusion of carbon.

The effect of the grain-size alone upon normality may easily be demonstrated. A steel should be selected which will, at some temperature such as 2100 degrees Fahr. (1150 degrees Cent.), develop in a brief heating a coarser austenite grain than at the carburizing temperature, e.g., 1725 degrees Fahr. (940 degrees Cent.) in the long period required to develop adequate case, 16 hours. At some time during the carburization one specimen is briefly withdrawn from the carburizing box and quickly heated to the coarsening temperature

Fig. 21—Influence of Austenitic Grain-Size Upon Structural Normality. Left—Coarsened by Heating Briefly to 2100 Degrees Fahr. (1150 Degrees Cent.) During Carburizing; Right—Carburized at Usual Temperature of 1725 Degrees Fahr. (940 Degrees Cent.). Both Cooled Together in Carburizing Box at Rate of 4.5 Degrees Fahr. Per Minute. × 1200.

and then when cooled to about the carburizing temperature replaced in the compound to restore the carbon lost during coarsening. The coarsened piece has the more normal structure, as shown in Fig. 21.

If an alloying element could be found which, as dissolved in austenite, raised the A_{e1} range materially, but at the same time retarded the reaction rate so little as to cause negligible increase in undercooling, then such an element would contribute directly to increased abnormality without altering the grain size. No such element has as yet been found. Aluminum, silicon and chromium, all of which raise the A_{e1} temperature so greatly, aggravate undercooling so that they are all contributors to normality, at least at cooling rates so far studied. Tungsten very nearly succeeds in accomplishing this but owing to its carbide forming propensities it so restricts grain growth as to fall instead into the broad category of vanadium. Aluminum, when dissolved and prevented from contributing alumina as an accelerating influence, is only a deep hardening element, like silicon. Oxygen does not serve in this capacity either, for it depresses the A_{e1} point.

Having then no alloying element which in solution contributes to abnormality, one has only the absence of alloying elements to rely upon to assist in the development of the abnormal structure from coarse austenite free from artificial stimulators of transformation, such as alumina or special carbide particles. However, the purest iron does actually permit structural abnormality to result from relatively coarse austenite. Carburized pure iron stands at the extreme end of a series of materials arranged in the order of increasing maximum grain-size compatible with an abnormal structure. Indeed, when one of the authors presented photomicrographs[3] of carburized pure iron with a normal structure he did not fully realize that the grain-size was so tremendously large as to make it incomparable with any ordinary material. As a matter of fact, pure iron-carbon steel can scarcely be rendered austenitic except with a coarse austenite grain. It is believed that a nearly pure iron-carbon alloy is the one material which, free from grain-growth inhibitors, may still show abnormality of structure when a grain-size in excess of about No. 1 is established. It possesses, at its high transformation temperature (minimum undercooling and high A_{e1} temperature), the maximum carbon diffusivity in relation to transformation velocity in the early stages.

As more and more alloying element—nickel, manganese, or silicon, etc.—is present the reversed state of affairs appears; structural normality is possible in finer and finer austenite grains as the alloy content increases; this is well exemplified in the fine-grained deep-hardening steels carrying upwards of one per cent manganese.

One factor influencing the degree of structural abnormality is now so well recognized and so completely in line with what has been related above that it scarcely warrants mention. The rate of cooling from the carburizing temperature in a measure determines the degree of ferrite divorce without, of course, changing grain-size in the slightest. Obviously, a high cooling rate increases undercooling and accordingly develops a more normal structure. Fortunately the rate of reaction is so very slow at only a moderate degree of undercooling and so rapid at a little greater undercooling that the actual cooling rate, over a wide range, changes the actual temperature at which the reaction begins only by a few degrees. Accordingly, for consistent results in normality studies, it suffices to use the same box and furnace without any special precautions to insure uniform cooling rate. Even different furnaces seem to yield essentially the

same structure. A good rate of cooling is 4 to 5 degrees Fahr. (2 to 3 degrees Cent.) per minute; a lower rate is satisfactory, a more rapid rate less desirable, although for purely comparative work one may more or less create his own standards.

The authors believe, in view of the above reasoning, that with an analysis at hand, together with a figure for grain-size as developed under the conditions of use, little if any further useful knowledge results from an observation of the structural normality or abnormality except perhaps in the most exceptional cases. At any rate no application for such information has been found not better served by other means. The information derived from a carburization test involving microscopic examination of the slowly-cooled piece is useful in estimating the effectiveness of the carburizing method and the grain-size established but the suitability of the steel is often more directly determined by carrying out the heat treatment and submitting the appropriate specimen to suitable hardness and other mechanical tests.

While it would be expected that the carburizing test would in any event be applied only to such steels as are to be carburized for service, it is not uncommon to carburize heat-treating steels and to evaluate grain-size and even structural normality therefrom. It should perhaps be stressed that such information is likely to be misleading unless the inapplicability of such data to the ordinary normalizing and other heat-treating operations is borne in mind. One may obviously not infer what the grain-size characteristics of a steel would be as heated, for example, 20 minutes at 1550 degrees Fahr. (845 degrees Cent.) from the grain-size developed after 8 hours at 1700 degrees Fahr. (925 degrees Cent.) in a carburizing environment. The great difference in time interval alone regardless of temperature differences would, in many steels, completely alter grain-size and hardenability beyond recognition. Apart from these obvious disagreements it should be borne in mind also that carburization is a process in which the constitution of a steel is changed. In the case of 1-inch rounds of carbon steel the hardenability may be changed even by a moderate carburizing followed by a subsequent heating for the comparison after quenching,—often in the direction of lower hardenability—occasionally in the opposite direction.

A striking example is to be found in a certain useful class of steels carrying a small amount of elemental aluminum left over after the deoxidation. One such steel as merely heated to about 1650 de-

grees Fahr. (900 degrees Cent.) was moderately coarse. When carburized (1725 degrees Fahr. or 940 degrees Cent. for 16 hours) it yielded an extremely fine grain in the extreme surface layer to a depth of about 0.016 inch, while the core and remaining hypereutectoid zone was coarse-grained. When reheated to 1725 degrees Fahr. (940 degrees Cent.) again, nearly the whole of the hypereutectoid zone was fine.

Clearly the carburizing test would have been most misleading in this case, even though the reasons are not hard to discover. The oxygen absorbed during carburizing combined with the aluminum and thereby provided grain growth inhibitors which could prevent the first small grains from growing. But the austenite grains were already coarsened in the interior before this oxygen penetration reached them; hence only a narrow outer zone of fine-grained material was found after cooling from the carburizing temperature. But in the coarse grains already established, the alumina continued to form in successively deeper zones as carburization continued. No particle dispersion, however effective, could make fine grains out of coarse ones already formed. However, upon the second heating they were effective in preventing coarsening and a deep fine-grained zone resulted. When this steel is heated briefly to the carburizing temperature, it acquires a coarse grain, for in the short time no oxygen penetration is possible and no grain growth inhibitors are formed.

Summary and Conclusions

For any ordinary medium or high carbon steel a correlation exists between the rate of transformation of the austenite to pearlite at any temperature and the grain-size of the austenite. It is thought that the greater grain boundary area of the fine-grained austenite provides more nuclei, or transformation centers, than exist in the austenite of a coarser grain structure; the degree of undercooling, at any specific cooling rate, is, therefore, less in the fine-grained austenite and the steel accordingly possesses a lower hardenability than if coarse.

Steels of the same composition may vary greatly with respect to the grain-size of the austenite developed at any particular heating temperature. The interest usually centers in the grain-size established at temperatures not more than two or three hundred degrees above the A_{c1} temperature and at this temperature some steels have

already coarsened while others remain fine. In general, a uniform grain-size is quickly established as soon as most of the carbide has dissolved. If this first uniform grain-size is large it has little potential capacity to coarsen further; if fine, it is likely to undergo an exaggerated grain-growth at some higher temperature.

The cause of the restraint of initial grain growth in some steels is believed to be the obstruction offered by dispersed particles of minute size—either non-metallic or special carbide compounds. The more effective these particles are, the higher will be the temperature required for marked grain growth in the austenitic matrix. These particles, apart from their controlling effect upon the grain-growth characteristics of a steel are thought to act in another capacity. They appear to act also as nucleation centers for the transformation of the austenite upon cooling. Thus, not only the actual small grains, but also the underlying cause of the small grains may tend to produce shallow-hardening steels.

The transformation products of fine-grained austenite, either martensite and its tempered derivatives or even pearlite, are somewhat tougher than those derived from coarse grains of austenite. This circumstance, discussed in greater detail elsewhere, is thought to relate to internal stress. Whatever the underlying cause, steels with inherently fine-grain tendencies are vastly superior for many applications as a result of their greater toughness.

The fine-grained steels are necessarily more shallow-hardening, other factors being equal; hence some alloying elements are incorporated to restore deeper hardenability when required. An example of the use of higher manganese to compensate in this direction for the desired fine grain is shown to illustrate the manner in which controlled grain-size operates.

Grain-size in hypoeutectoid and hypereutectoid steels may be estimated microscopically in specimens prepared by cooling in such a manner as to produce first a pro-eutectoid network and then cooling thereafter somewhat rapidly to produce a dark-etching matrix in the body of the parent grains. Eutectoid steels should be quenched just more slowly than the critical quenching rate, so as to mark out the grain boundaries with dark-etching fine pearlite with martensite constituting the body of the parent grains. A special case of grain-size estimation is that of case carburized steels in which the slowly cooled hypereutectoid zone is ideally constituted for the purpose. When applied to low carbon, low alloy carburizing steels certain character-

istics of structure serve to classify the steels of high and low hardenability. To these extremes of structure the names "normal" and "abnormal" have been applied. The immediate cause of the abnormal structure is high carbon diffusivity relatively to initial reaction rate. Normal steels are those having coarse grain-size or carrying higher proportions of alloy or both. Abnormality is most pronounced in nearly pure iron-carbon steels having fine grain-size.

DISCUSSION

Written Discussion: By H. S. Rawdon, Chief, Metallurgical Department, National Bureau of Standards, Washington, D. C.

The writer had occasion recently to review the metallurgical accomplishments of the past few years and to give his appraisal of them as to their relative importance, actual or potential. In the field of carbon steels the establishment of the incontrovertible fact that steels of same composition, as ordinarily determined, can differ decidedly in their intrinsic properties, particularly those in which heating is involved, is without doubt the outstanding accomplishment. Vague, indefinite claims of this kind have been made in past years, of course, but the work of McQuaid and Ehn was the first to give such claims any experimental support. Later work such as that summarized at this meeting by Davenport and Bain and by Epstein and his associates in their respective papers serves admirably to clarify the situation still further and to show wherein well established metallurgical principles apply. In view of the information presented at this meeting, it is difficult to imagine that any "doubting Thomases" can still remain.

One cannot help but ponder on the possibility that perhaps results reported for carbon steels in past investigations may have been materially influenced, entirely unwittingly on the part of the investigator, by this "quality" variable as a result of the deoxidation practice used. A case of this kind which recently came forcibly to the writer's attention arose in connection with the work of the Joint Research Committee on the Effect of Phosphorus and Sulphur on Steel —an investigation which has been under way for many years. As basis for the study of the effect of phosphorus on low carbon steel used in pipe manufacture, four special heats of steel were made in 1928 by one of the well-known steel pipe manufacturers. The special inspection committee having supervision of these special heats, the composition of which was varied so as to give phosphorus progressively increasing up to 0.10 per cent, complacently congratulated each other on the success attained in their job when they reviewed the mill analysis of the steels. According to the information then available, there was no reason whatever to believe otherwise than that such differences as might be found later in the strength properties in testing these steels could properly be attributed to the arbitrarily introduced differences in the phosphorus content.

However, on carrying out the mechanical tests, it was found that one heat was decidedly out of line with the others particularly in its impact resistance. Quenching served to emphasize this difference. Fig. 1 (of this discussion) is typical of the difference in impact resistance. Examination of the microstructure

indicated that the heat in question was of "abnormal" steel (Fig. 2) and subsequent examination of the mill records showed wherein the deoxidation treatment would account for this. In this particular case, therefore, this so-called "quality" variable, the introduction of which was unintentional and unsuspected, entirely obscured the effect which was expected to result from the increased

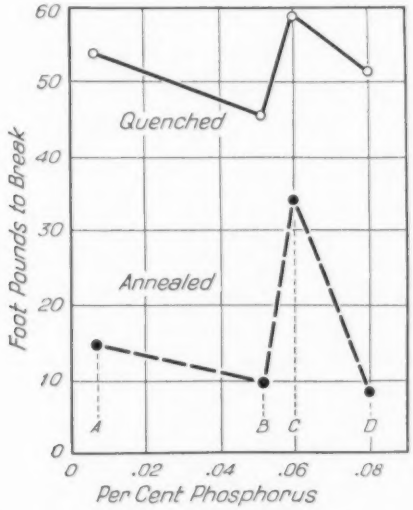


Fig. 1—Impact Resistance (Charpy) of Longitudinal Specimens of Low Carbon Steels of Progressively Increasing Phosphorus Content. The Composition in Per Cent was as Follows:

Heat	P	C	S	Mn	Si	Ni	Cu	Cr
A	0.007	0.10	0.029	0.47	0.01	0.016	0.02	0.005
B	0.051	0.10	0.030	0.49	0.02	nil	0.04	0.009
C	0.060	0.11	0.030	0.48	0.03	0.026	0.08	0.010
D	0.080	0.12	0.029	0.48	0.03	nil	0.02	0.014

phosphorus content. Other similar cases of this kind undoubtedly exist. The necessity for still more closely defining any steel which is to be used for test purposes is very evident.

The lack of a suitable nomenclature for use in discussing this subject is deplorable and is well illustrated by papers at this meeting. Even now that the basic underlying principles are fairly well understood, authors find it very necessary to define the terms which are to be used as is well illustrated by the paper of Messrs. Davenport and Bain. Thus, they make it clear that though this is a "grain-size" phenomenon it is not grain-size in the sense that this term is generally accepted. The American Society for Testing Materials grain-size chart for steels which has been referred to in a number of papers at this meeting is in no sense analogous or comparable to the similar grain-size chart adopted by the same organization for nonferrous metals such as copper and alpha-brass. The time has certainly come for concerted action on this matter of nomenclature. The American Society for Metals has unquestionably done most in this particular subject and it would seem entirely proper for the Society to take the lead in the matter. The appointment of a small committee

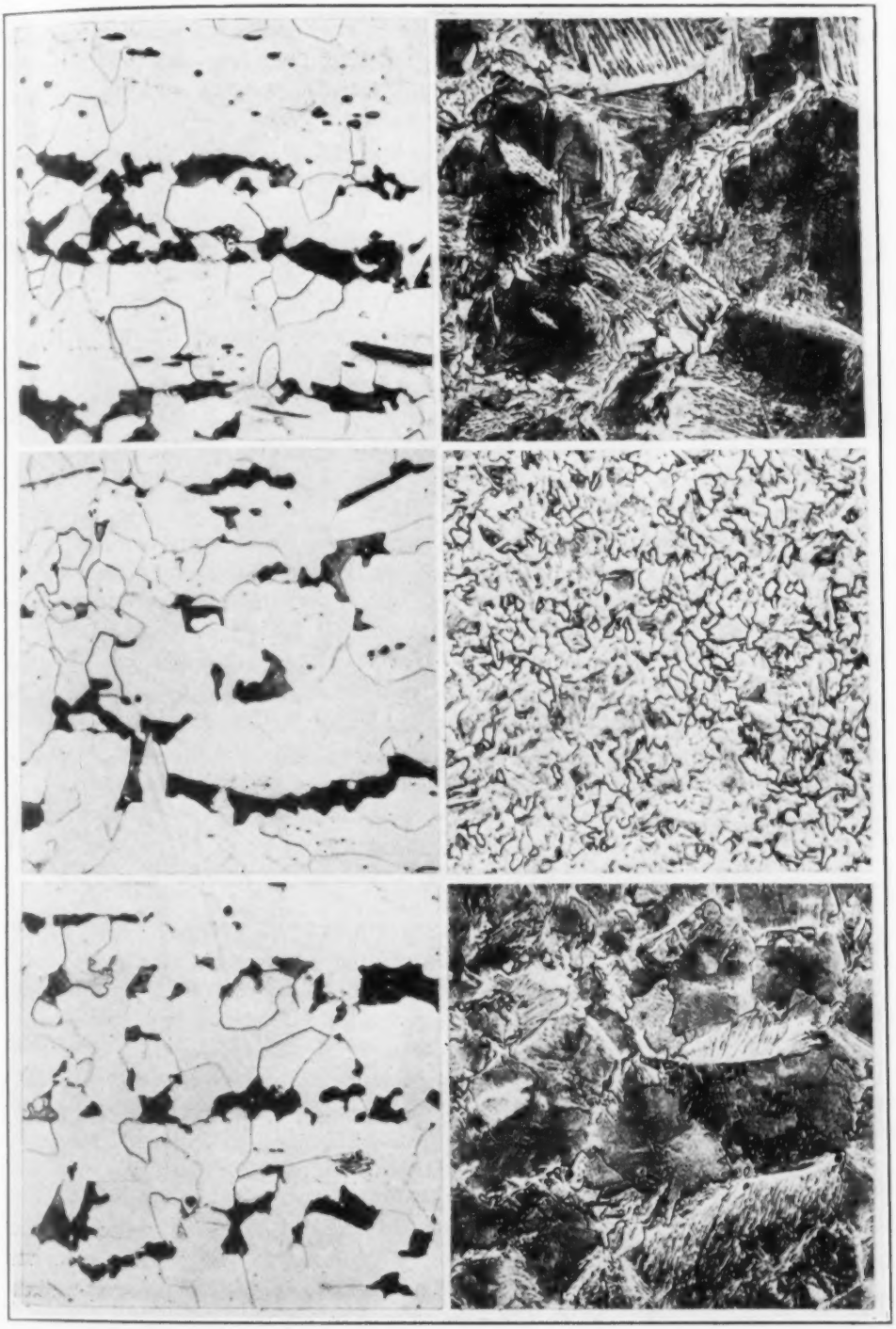


Fig. 2—Microstructure of Three Heats of Low Carbon Steel to Which Phosphorus was Added in Progressively Increasing Amounts; Etched in 1 Per Cent Alcoholic Nitric Acid; Initial Magnification, × 100, Reduced. Photomicrographs on Left Show the Annealed Steel (1725 Degrees Fahr., Cooled in Furnace), Those on the Right the Same After Water Quenching (1725 Degrees Fahr.). The Composition is Given in Fig. 1, Upper Row of Photomicrographs, Heat B; Center Row, Heat C; and Lower Row, Heat D.

to co-operate with other technical societies as was done in the case of heat treatment terms would seem to be in order. To the writer, it would seem that a relatively simple precaution should go far in clarifying the situation. In brief, this is always to use the specific term "austenite grain-size" in speaking of the subject of quality in steel and not simply "grain-size." This is what Davenport and Bain have done and the writer regrets that it does not appear in the title of their paper. Austenite is now a well-known term to metallurgists and no one need fear being dubbed a "high brow" if he uses it in specifying steel, the term "controlled austenite grain-size" instead of simply "controlled grain-size."

B. F. SHEPHERD:[1] The authors are to be congratulated very much on this contribution. Any contributions by Davenport and Bain are looked forward to with a great deal of interest. There are two points I would like to bring up —one is in regard to Fig. 14, in which a comparison of coarse-grained steels makes no mention of the prehardening structure, whereas practically all of the other experiments state what the prehardening structure was. We have found this information to be quite essential as having a marked influence on the results obtained in the final hardening.

There is another rather minor point of technical procedure. We make quite a number of hardness penetration curves on tool steel by Rockwell and Vickers methods, and have had to take average results of the hardness tests at any distance in from the center. It appears that any curve along any one radius would vary quite a little from a curve along another radius. A great many of the curves presented by the authors show the hardness penetration from one side to the other side of the specimens to be what might be called identical twins. I am much interested to know whether these curves were plotted that way for the sake of illustration from one set of readings, or whether the readings actually did fall on the line with that degree of uniformity.

Authors' Reply

In answer to Mr. Shepherd's question as to the prehardening structure of the specimens of Fig. 14, we would say that it is our usual practice, in hardenability determinations of this nature, to subject the material to a pre-treatment designed to eliminate carbide segregations and other variables which might exist in the "as received" steel. This pre-treatment usually consists of heating the "as-received" material to a temperature sufficiently high to take practically all of the carbide into solution, oil quenching and then tempering to render the material soft enough for machining to size for final hardening. This leaves the carbides in a finely divided condition—either fine pearlitic, or finely spheroidized, or both—readily soluble in the austenite on the subsequent heating for hardening. For example, the steels of Fig. 14 were oil-quenched from 1600 degrees Fahr. (870 degrees Cent.) in the form of 1⅛-inch diameter rounds, tempered at 1300 degrees Fahr. (705 degrees Cent.) for 30 minutes at heat and then machined to the final size of 1-inch diameter.

In regard to Mr. Shepherd's other question on the plotting of the hardness-penetration curves, it is our practice to take the Rockwell hardness readings on

[1]Manager, Rock Drill Division of the Ingersoll-Rand Co., Phillipsburg, New Jersey.

a series of concentric circles on the transverse section of the hardened specimen. Four or more readings are taken on each circle, and all of the readings on any one circle are then averaged. Surface readings are taken after removal of decarburized metal and also a center reading. The average of all the readings on any one concentric circle is plotted on each side of a central vertical line in the hardness-penetration diagram, the distance from the mid-vertical being proportional to the actual distance involved on the specimen. This, of course, makes the two halves of the diagram mirror images of each other. We have felt that this method of plotting gives a more useful and practical description of the trend of hardness penetration than any diagram involving all of the minor and irrelevant variations which may occur along any single, chosen diameter.

The case, cited by Dr. Rawdon, of a steel, the anomalous behavior of which could not be elucidated on the basis of ordinary chemical analysis, is an excellent example of the influence of grain-size and deoxidation practice upon steel quality. We venture to say that this problem could not have been analyzed and the behaviors accounted for without our present knowledge of grain-size effects. Incidentally, it is interesting to note that the grain-size evidence presented by Dr. Rawdon was apparently secured without recourse to the conventional carburizing test, and yet the grain-size implications of the problem are as clear-cut as anyone could wish. All of which lends support to the argument for more "unadulterated" grain-size information and less blind adherence to the traditional carburizing test in our efforts to arrive at the fundamentals of this particular phase of steel behavior.

We are entirely in agreement with the sentiments expressed by Dr. Rawdon in his remarks on nomenclature. The sooner a rational grain-size terminology is established—and this applies equally well to certain other branches of steel metallurgy—the sooner will we be rid of much of the confusion that, unfortunately, still seems to exist on this subject.

GRAIN-SIZE IN RELATION TO MACHINABILITY AND OTHER PROPERTIES OF BESSEMER SCREW STEEL

By H. W. Graham

Abstract

There are presented data obtained in a study of fourteen lots of commercial cold drawn Bessemer screw steel. The chemical and physical properties are compared with the machinability of the various lots and with grain counts in the "as received" condition and as shown by the McQuaid-Ehn test. No clear correlation on the lots studied is shown between machinability and either chemical analysis or grain count by the methods used.

Data are shown on the work brittleness and work hardness characteristics together with some exploration of the effect of accelerated aging, and the technique by which the tests were made is described.

With no clear correlation established between machinability and grain-size, it is argued that no estimate of the effect of grain-size on machinability can be obtained until some information is developed to show the relative energy expended in rupturing a grain boundary as compared to the energy required when the path of rupture passes through the grain itself.

IT IS recognized that grain-size and quality have been studied to the greatest extent in the field of carburized parts and in the manufacture of heat treated articles made from forging steel. In both cases, grain-size has been considered to have a relationship of some importance to the machinability of the steel. Since in Bessemer screw steel (which is frequently carburized) we are dealing at first hand with the matter of machinability; it does not appear that the subject matter presented in this paper is unrelated with other work on the subject of grain-size and quality.

We do not wish to argue that any conclusions presented in this paper on the machinability of Bessemer screw steel would necessarily apply on forging steel, in heat treated gears or other similar applications. However, it may be that if there exists a positive and effective correlation between grain-size and machinability on forging steel,

A paper presented as part of the Grain-Size Symposium of the Sixteenth Annual Convention of the Society held in New York City the week of October 1, 1934. The author, H. W. Graham, a member of the Society, is general metallurgist, Jones and Laughlin Steel Corp., Pittsburgh. Manuscript received July 2, 1934.

before or after heat treating, then some evidence of a similar correlation might be found in the automatic machining of Bessemer screw steel. This is, however, a theorem which we shall not attempt to develop.

The data presented consist of observations on a series of fourteen sample lots of Bessemer screw steel in ¾-inch cold-drawn rounds, representing the product of six different commercial suppliers. The samples were obtained in a manner which guarantees that at least four different Bessemer steel-making plants are represented. There was approximately one ton of round bars in each lot. The six sources of supply are referred to in this paper under the code letters A. B. etc., while different lots are listed as A1, A2, etc. The circumstances of the sampling were such as to almost certainly guarantee that the fourteen lots represented fourteen different heats of steel. The chemical analyses of the materials are shown in Table I.

It is to be noted that lots B3, D2 and E1 represent the so-called "high-sulphur" screw stock. Since we are to consider grain-size, as well as machinability, the analyses of silicon and aluminum have been included. These elements in various forms may well act to inhibit grain growth and may, therefore, have an effect upon grain-size.

An index of machinability in Tables I, II. III is shown with the analyses in the form of tool-life in minutes. The machinability test was made in a standard four-spindle automatic lathe, with no lubrication on the tool. The figures on tool-life are for a surface speed of 234 feet per minute and a feed of 0.0085 inches, although the data for other feeds and speeds are fully given later in this paper. When carefully performed the dry machining test is fairly satisfactory, particularly on Bessemer screw steel which shows a quite definite end point. Since the tool-life shown is for dry-cutting, the actual figures given will not apply to commercial shop operation with proper tool coolant. The ratio between tool-life dry and with proper cooling varies considerably. At unusually high speeds or heavy feeds the tool-life with a coolant may not be more than twice that when the tool is run dry; at lower speeds and feeds the ratio may increase to 4 to 10 times the dry tool-life. The physical properties of the cold-drawn bars are shown in Table II.

The data on grain-size are shown in Table III. The pearlitic grain count was made on the cold-drawn bars in the "as-received"

Table I

Chemical Composition and Tool-Life

Code	Carbon	Manganese	Phosphorus	Sulphur	Silicon	Al_2O_3	Tool-Life in Minutes
A1	0.10	0.65	0.093	0.181	0.020	0.0019	97
A2	0.10	0.70	0.082	0.161	0.019	0.0010	100
B1	0.14	0.87	0.115	0.175	0.017	0.0048	45
B2	0.11	0.82	0.102	0.161	0.013	0.0094	85
B3	0.13	0.84	0.110	0.266	0.020	0.0010	123
C1	0.11	0.80	0.104	0.190	0.026	0.0004	41
C2	0.11	0.82	0.105	0.189	0.033	0.0027	42
C3	0.11	0.80	0.107	0.208	0.017	0.0029	88
D1	0.11	0.83	0.101	0.181	0.020	0.0030	54
D2	0.10	0.90	0.086	0.263	0.025	0.0030	45
E1	0.11	0.93	0.098	0.248	0.020	0.0048	45
F1	0.10	0.78	0.091	0.166	0.027	0.0030	97
F2	0.11	0.84	0.107	0.179	0.018	0.0030	74
F3	0.11	0.84	0.114	0.169	0.012	0.0004	100

Table II

Physical Properties and Tool-Life

Code	Yield Point Pounds per	Tensile Strength Square Inch	Elongation in 2 Inches	Reduction of Area	Brinell	Izod	Tool-Life in Minutes
A1	89,200	93,100	19.0	43.3	174	15	97
A2	86,890	91,090	20.5	50.7	179	19	100
B1	81,970	91,560	23.0	49.8	187	12	45
B2	85,220	90,960	23.5	48.4	179	19	85
B3	87,840	92,280	19.5	40.8	183	16	123
C1	88,550	95,260	22.0	47.7	179	12	41
C2	93,680	98,230	22.0	47.5	196	6	42
C3	85,710	91,310	21.0	42.4	174	25	88
D1	95,200	99,890	20.5	43.8	192	16	54
D2	92,040	95,800	19.0	44.2	187	14	45
E1	98,540	100,250	17.0	39.6	202	16	45
F1	91,460	96,360	19.5	48.7	183	27	97
F2	94,870	97,470	20.0	46.8	192	15	74
F3	92,990	95,930	19.0	44.3	179	9	100

Table III

Grain-Size of Cold Drawn and Carburized Specimens and Tool-Life

Code	Cold Drawn	Hyper-Eutectoid	Hypo-Eutectoid	Core	Normality	Tool-Life in Minutes
A1	214	6	12	22	A	97
A2	140	7	20	33	A	100
B1	126	7	13	26	N	45
B2	150	8	16	25	PA	85
B3	166	8	14	24	A	123
C1	157	11	21	29	PA	41
C2	197	9	17	37	PA	42
C3	169	12	21	29	A	88
D1	274	10	12	22	PA	54
D2	307	11	20	32	A	45
E1	166	10	26	29	A	45
F1	168	19	23	31	A	97
F2	206	9	13	24	A	74
F3	148	7	13	28	PA	100

A—Abnormal; N—Normal; PA—Partially Abnormal

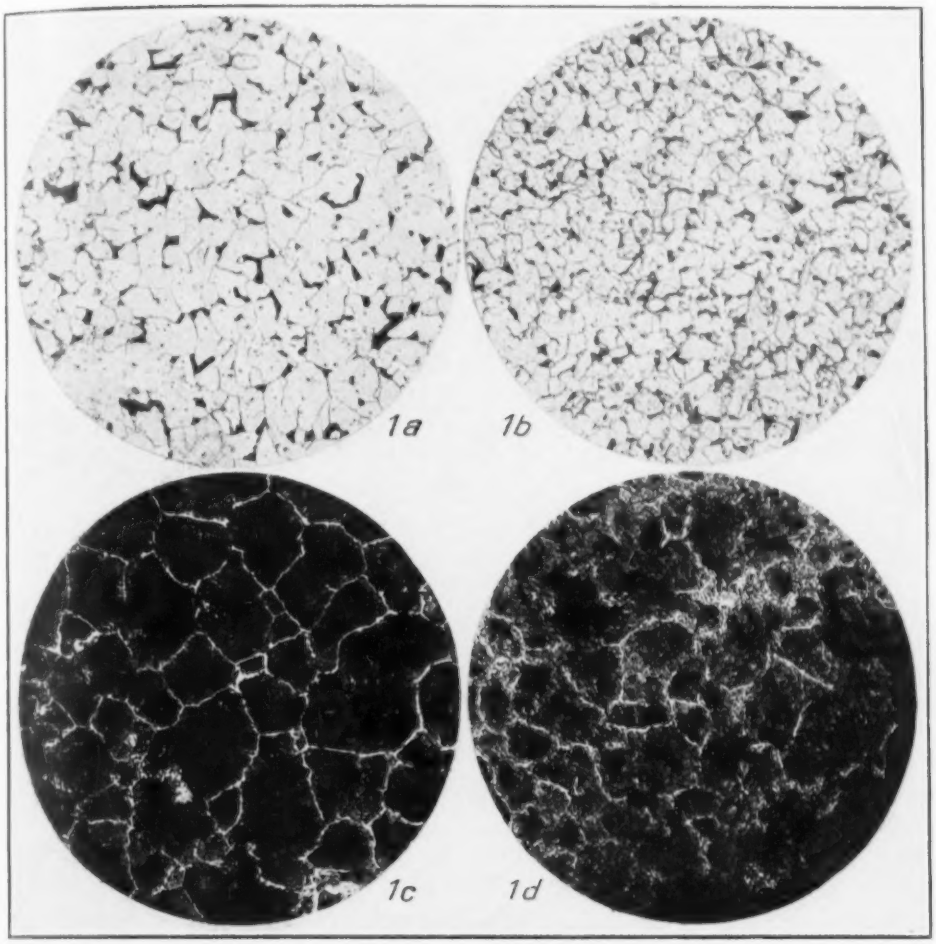


Fig. 1a—Code A2, Large Grain Type, Cold Drawn Condition.
Fig. 1b—Code D2, Small Grain Type, Cold Drawn Condition.
Fig. 1c—Code B1, Normal Type, McQuaid-Ehn Test.
Fig. 1d—Code E1, Abnormal Type, McQuaid-Ehn Test. All Photomicrographs × 100.

condition. The austenitic count was made after the specimens had been subjected to the standard McQuaid-Ehn carburizing test at 1700 degrees Fahr. (955 degrees Cent.) for eight hours.

With respect to the pearlitic grain-size in the cold-drawn condition, while all specimens are to be listed in A. S. T. M. Class 8, there is still a considerable range of size from 126 in specimen B1 to 307 in specimen D2, as shown in the photomicrographs Fig.1 a and b.

In Table III the specimens are listed as to normality. Photomicrographs Fig. 1 c and d illustrate the range of normality. The specimens from source B showed tendency toward normality, but

aside from this all specimens were clearly toward the abnormal end of the range.

A machining speed of 234 feet per minute is approximately four feet per second, which is a sufficiently rapid rate to be properly classed as dynamic. Data will, therefore, be presented relative to the physical performance of the test materials under the deforming action of machining. This phase of the investigation was approached by

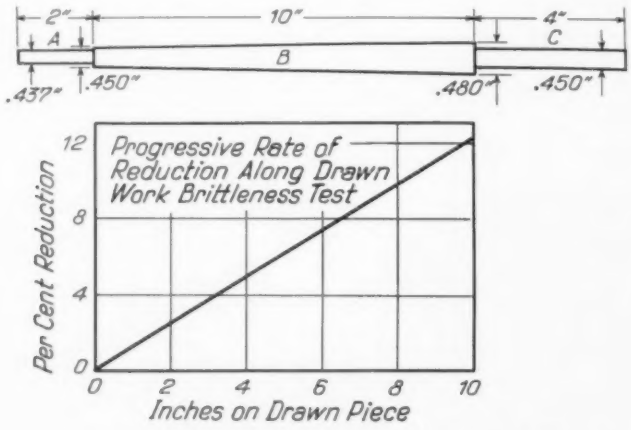


Fig. 2—Standard Work-Brittleness Test Bar and Curve for Progressive Rate of Elongation of Drawn Work-Brittleness Specimens.

the "work brittleness" test that has become a standard method of procedure in our organization. A tapered test-piece is normalized and then machined according to the dimensions shown in Fig. 2 and then cold-drawn through a 0.450-inch die.

It is obvious from the shape of the test-piece that after drawing it will be in a progressively cold-worked condition along its length. The test-piece is ordinarily subjected to accelerated aging by heating to 550 degrees Fahr. (290 degrees Cent.) for one hour. Then the test piece may be properly notched for the Izod test at successive points along its length, thus determining the loss of impact resistance with increasing cold-work. By this method were determined the data presented in Table IV.

If the test-piece is tested for hardness progressively along its length before the Izod notches are machined, the rate of increasing hardness with increasing cold-work will be established. By this means were determined the figures given in Table V.

The data given in Table V show the hardness values for a test-piece in the unaged condition. If the test-piece is subjected to accel-

Table IV

Work Brittleness Tests

Code	Unstressed	Izod Figures Foot-Pounds Per Cent Reduction in Cold Drawing .5	2.3	4.4	6.3	8.0	10.0	12.0
A1	(14-20)	16	3	2	2	2	2	2
A2	(78-76)	84	16	4	3	2	3	1
B1	(50-56)	34	18	2	2	2	1	1
B2	(78-75)	35	2	2	2	2	4	2
B3	(46-70)	27	5	2	2	2	2	2
C1	(34-71)	38	5	2	2	2	2	2
C2	(80-80)	20	4	2	2	3	3	2
C3	(76-58)	58	17	4	2	3	2	2
D1	(70-70)	71	62	19	16	2	2	2
D2	(72-72)	70	26	20	3	8	4	2
E1	(70-76)	70	10	2	2	2	3	2
F1	(75-82)	80	80	4	4	4	2	3
F2	(70-75)	25	3	2	2	2	2	2
F3	(25-10)	12	2	2	2	2	2	2

Table V

Work Hardness Tests Unaged. Rockwell "B" Scale

Code	Unstressed	Per Cent Reduction in Cold Drawing .5	1.4	3.4	5.4	7.2	9.0	11.0	12.0
A1	67	62	71	78	79	80	81	82	81
A2	64	61	70	79	82	84	86	84	87
B1	75	78	79	84	93	93	93	94	95
B2	69	72	83	90	94	93	93	94	94
B3	68	69	72	84	88	90	90	91	90
C1	71	73	74	82	90	90	93	91	92
C2	71	73	78	82	87	89	89	89	89
C3	69	66	71	77	80	82	82	84	83
D1	70	75	78	85	87	93	94	93	93
D2	69	76	79	83	84	92	93	92	93
E1	68	68	72	80	83	86	85	86	86
F1	63	64	69	79	85	87	87	89	89
F2	65	65	73	77	79	81	81	82	81
F3	72	71	78	84	89	93	93	94	94

Table VI

Work Hardness Tests Aged. Rockwell "B" Scale

Code	Per Cent Reduction in Cold Drawing .5	1.4	3.4	5.4	7.2	9.0	11.0	12.0
A1	73	79	84	84	90	84	86	89
A2	61	65	78	80	84	84	86	87
B1	77	78	82	91	92	93	93	93
B2	74	78	86	90	91	92	91	92
B3	68	73	84	87	92	87	93	93
C1	71	77	84	88	90	90	90	92
C2	74	79	86	88	95	90	90	95
C3	68	70	88	86	87	92	91	92
D1	75	77	82	85	92	92	92	93
D2	75	78	80	82	91	91	92	92
E1	64	74	85	89	89	88	88	88
F1	62	75	85	92	86	86	91	89
F2	64	78	79	86	93	94	95	89
F3	71	79	83	89	94	94	93	94

erated aging for one hour at 550 degrees Fahr. the hardnesses are as shown in Table VI.

As a final step in presentation of the data observed in this investigation, there is shown in Fig. 3 the machinability performance for some of the various lots of screw steel included in this investigation. Part of the tool-life curves have been omitted to make a clearer chart.

The curves show the relationship of the speed of machining to tool-life. The machinability might have been shown by depth of

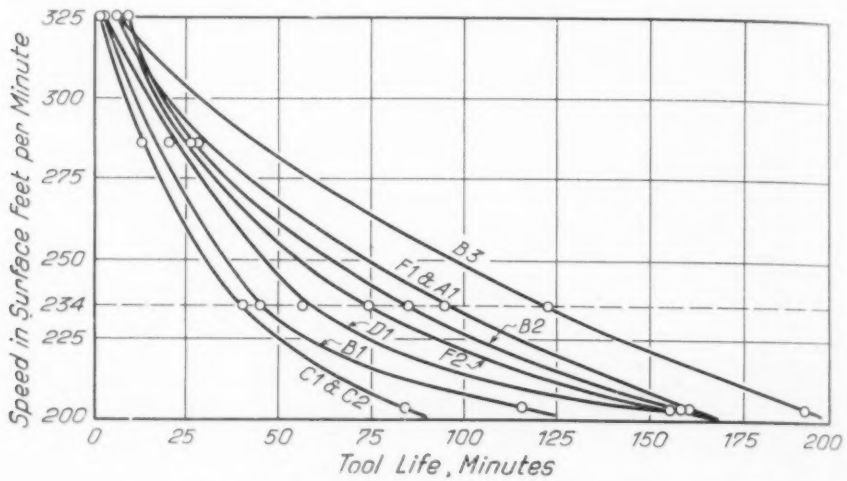


Fig. 3—Machinability Tests Showing Speed Versus Tool Life.

cut plotted against tool-life or by volume of metal removed per unit of time plotted against tool-life. The shape of the curves obtained would be slightly different, but there would be no real difference in the showing of one lot as compared to another.

The horizontal dashed line in Fig. 3 shows that the tool-life figures used in this paper as an index of machinability were selected at 234 surface feet per minute. This speed was selected as representing a high figure for commercial operation, but it is to be noted that the relationship of one steel to another is about the same at any speed. This is an important point because it indicates clearly that a lot of steel which machines poorly at any one speed may be expected to perform poorly at any speed.

Turning our attention to Table III with its data on grain-size, we will first examine the pearlitic or "as-received" cold-drawn grain-size compared with tool-life. The values when plotted are as shown in Fig. 4. Although no line is drawn in Fig. 4 to define a relationship

between grain-size and machinability, yet such a line would show an improvement of machinability with an increase of grain-size.

When it comes to the question of austenitic grain-size we will not attempt to present any correlation with machinability, because the examination of the data does not appear to present any clear case for such correlation, except a very slight suggestion that better machinability is to be expected with larger austenitic grain-size.

We recognize that this investigation might well include a study of the variation that might be found as to the temperature at which

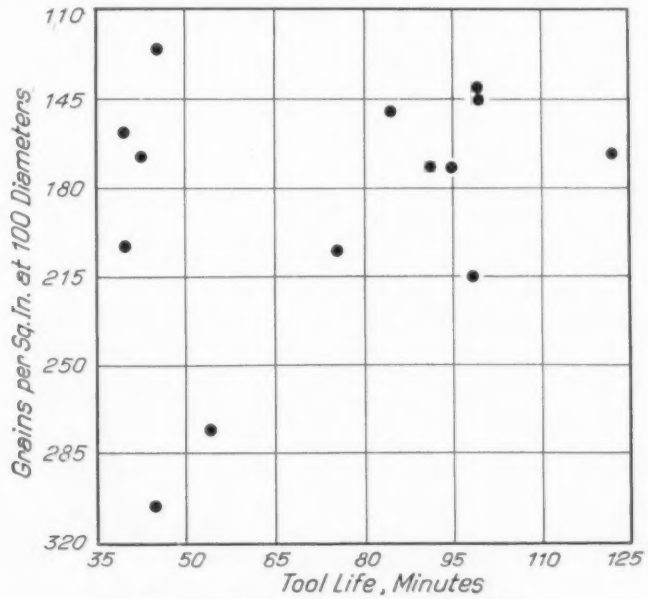


Fig. 4—Grain-Size Versus Tool Life.

grain coarsening takes place. After some deliberation we have decided not to extend this paper to the length necessary to properly present this phase of the subject.

Examination of the figures shown in Table IV discloses a sharp loss of impact strength with increasing increments of cold-work. The degree of separation between various lots shows this test to be highly discriminating. This fact has caused it to be included in our standard testing procedure ever since its origin in our laboratories some six years ago.

In Fig. 5 are plotted curves showing the loss of impact resistance with increasing cold work for a few representative cases of the fourteen lots hereunder investigation. The range of behavior is striking, but so far as this investigation is concerned it must be said that there is no clear correlation between machinability or grain-size and

sensitivity to cold-work. Data in our possession beyond the range of this investigation, however, show that in general best machinability is to be expected where there is the greatest loss of impact

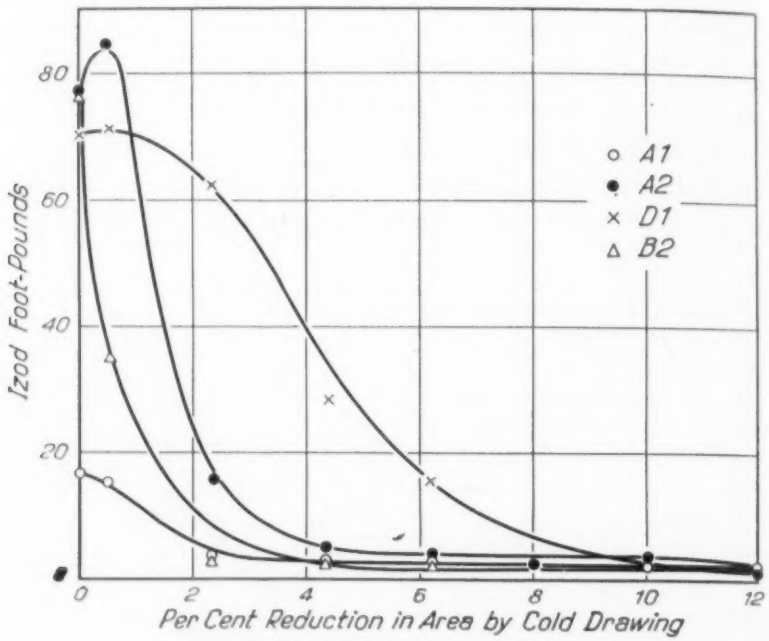


Fig. 5—Work-Brittleness Test Curves.

strength in the work-brittleness test. The fact that the correlation with machinability is not more definite is due to a number of factors, some of which are more or less debatable in nature.

The degree of cold-work shown in Fig. 5 extends from the normalized condition to a maximum of 12 per cent reduction. This figure is representative of commercial cold-drawing, and we might, therefore, say that this test shows the progression from the hot-rolled (more accurately the normalized) condition to the cold-drawn condition. The data given in Table IV, however, are for specimens that have been subjected to accelerated thermal aging, and, therefore, the impact value at 12 per cent reduction in Table IV may be expected to be lower than in commercial cold-drawn Bessemer screw steel, since commercial screw steel is not fully aged. Commercial cold-drawn screw steel does, under prolonged natural aging, tend to fall to the values here given in the sensitivity test. Natural aging is undoubtedly influential in improving machining performance. The degree of this effect varies from one heat to another, apparently due to a variation in the aging capacity of different heats.

In discussing the failure of the work-brittleness data here given to agree closely with the figures on machinability, some attention may be given to the extent to which the Izod impact test does or does not correctly simulate the cutting action of a lathe tool. There appear to be grounds for concluding that the action is quite similar. In many open-hearth grades of steel the impact test is distinctly of a bending nature. In the field of cold-drawn Bessemer screw steel it includes a combination of bending and breaking which appears quite similar to the action of an automatic screw machine tool cutting at high speed. A low Izod figure certainly displays the characteristic in the steel of a localized stress at the notch, and this must be associated with good machinability. When the Izod impact strength creeps up above twenty foot-pounds, it shows conclusively that the nature of the steel is such that stresses are dissipated throughout a greater volume about the notch. This means that from a machining standpoint that the tool must perform its work in this greater volume with a greater total expenditure of energy. Consequently a greater amount of heat is developed, resulting in shorter tool-life, more frequent interruptions, decreased production, and poorer surfaces and poorer dimensional accuracy. All of these are a part of poor machinability.

One factor has, however, been omitted from our discussion of impact properties as related to machinability. The data given in Table IV represent impact tests performed at room temperatures. Obviously the tool which cuts Bessemer screw steel is operating at a somewhat elevated temperature. To explore this point we have conducted impact-shear tests through a range from room temperature to 600 degrees Fahr. (315 degrees Cent.), which should cover the machining temperature in all cases. These data, however, are not reported in this paper.

As a first step the test-pieces were subjected to a uniform normalizing for one hour at 1675 degrees Fahr. (915 degrees Cent.). While this may be superficially accepted as having brought the samples to a uniform and unstrained condition, yet there is some justification for the thought that each test specimen should have been normalized at or above that particular temperature which is critical with respect to grain coarsening. Such treatment might be expected to have some bearing upon further behavior of the specimen in the sensitivity test or in other investigations.

As a matter of technical and commercial interest, it should be mentioned that there is a considerable difference in behavior with respect to embrittlement and hardening when steel is cold-worked by cold-drawing, by cold-rolling, or by some compressive method. In compression the least difference is developed, while in cold-rolling

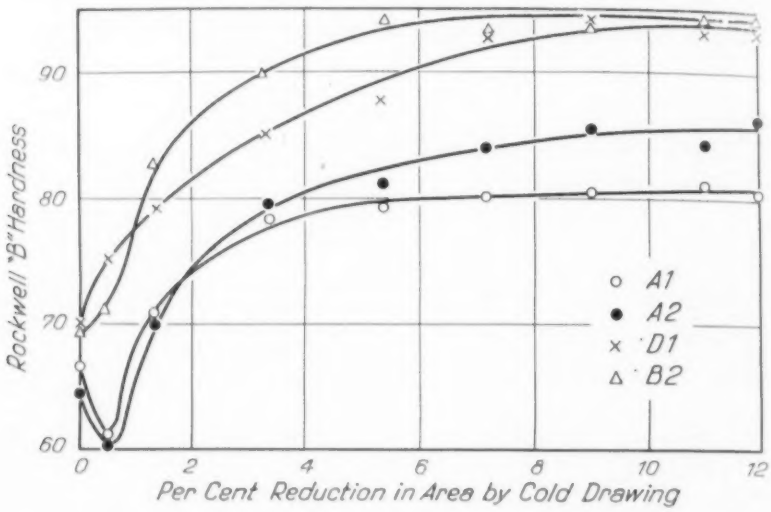


Fig. 6—Work-Hardness Tests of Unaged Specimens.

there is the greatest loss of impact strength. This reaction appears to be tied in with an improvement of fatigue properties, which would explain the splendid fatigue performance of cold-rolled shafting and cold-rolled axles.

Consideration of the loss of impact strength should be carried on coincident with attention to the matter of increase of hardness with increasing cold-work. Data relative to work-hardness are shown in Table V for unaged specimens. As in work-brittleness, the increase of hardness proceeds concurrently with the loss of impact strength, as shown in Fig. 6.

If the specimen is tested for hardness throughout its length before and after the thermal aging, the effect of the thermal aging is made apparent. In an occasional case there is little or no increase in hardness in thermal aging, but in most cases there is a considerable aging effect due to the thermal treatment. Fig. 7 shows two cases in which there is a wide variation in the effect of thermal aging. It can hardly be argued that such a variation can be a function of grain-size, except in a very indirect manner. Presumably this variation is a function of varying quality of the ferrite.

We have found that variations in testing procedure either mechanical or thermal may produce quite a difference in the increase of hardness with little or no change in loss of impact strength. It is, therefore, important to note that *work-hardness and work-brittleness do not proceed in fixed relationship to each other.*

No satisfactory correlation is indicated between the work-hardness data here shown and machinability. While it may be taken

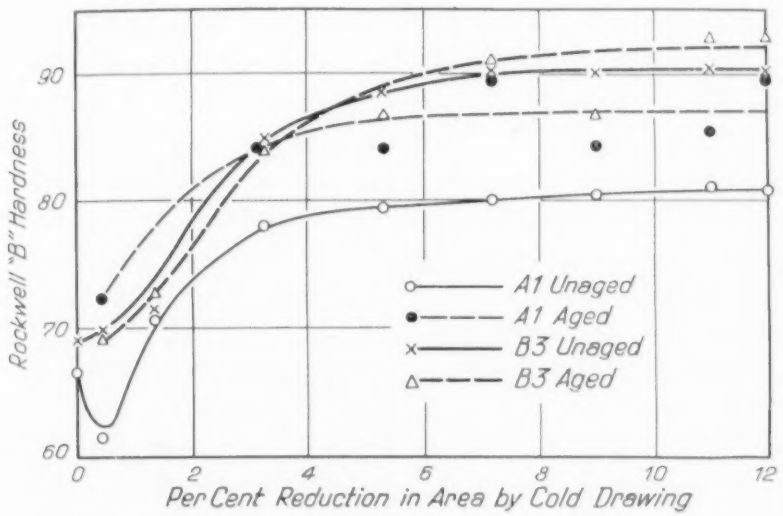


Fig. 7—Work-Hardness Test Curves. Circles and Crosses are of Unaged Specimens. Dots and Triangles are of Aged Specimens.

as a conclusion in a generally correct direction that best machinability is associated with maximum work-brittleness and minimum work-hardness, yet there are other considerations which must be borne in mind. It might be assumed that the minimum energy required for rupture would be associated with minimum hardness, yet practical experience shows that very soft steels are notoriously difficult to machine. An example of this is S.A.E. 1015 steel. With increased hardness there is a decreased "mushing" action; that is, the cutting force penetrates a lesser distance into the metal around the tip of the tool. Less volume of metal is deformed in the process of rupture and, although the amount of energy which must be expended per unit volume of metal is increased due to the increased hardness, yet the total energy is diminished by reason of the lesser volume of metal involved. The decrease in volume becomes less with increased hardness until a point is reached beyond which there is no further decrease in the total amount of metal involved in the

cutting operation. This is the point of maximum machinability; that is, of minimum total energy required. Beyond this point any increase of hardness is accompanied by a proportionately greater total energy which means a corresponding decrease in free-cutting qualities.

The reasoning here presented brings us to what must be, for the purpose of this paper, our final position on the subject of grain-size and machinability.

It must be admitted that the data presented in this paper do not make clear any definite relationship between grain-size and machinability. From a practical point of view we must assume that this is due to an insufficient number of test lots and to the fact that the various lots have not been reduced to a condition of proper comparability. However, it is not likely that any satisfactory demonstration of this relationship will ever be produced by the metallurgist. The answer must be presumed to lie in the field of physical metallurgy and it is to the physicist to whom we must look for help.

Those who are accustomed to consider either that grain-size is of great importance as related to machinability, or that grain-size is at least an important indication of machinability, are accustomed to take the position that in the cutting operation the path of rupture follows preferred planes along which the separation proceeds rather easily, and that the grain boundaries interrupt the progress of the line of separation along these preferred planes. While the correctness of this point of view is not questioned, yet the degree of importance which is attached to it must depend upon the relative energy expended in forcing the line of separation along the preferred planes through the crystal as compared to the amount of energy expended at a grain boundary. From our own point of view, while we admit that the grain boundary must necessarily involve some additional expenditure of energy, yet we find it difficult to believe that the amount of additional energy expended at the grain boundary is very large as compared with the normal energy required in forcing the path of rupture through the crystal itself. If no energy were required to force the path of rupture through the crystal and if some energy were required at the grain boundary, then machinability would be strictly proportional to the number of grains or to the grain-size. If, at the other extreme, the additional energy necessary to cross the grain boundary is only a very small proportion of that required to cross the crystal, then obviously grain size is of very little importance as concerns machinability. There is reason to think that

the energy required within the mass of the crystal itself is of a considerable order, in spite of preferred planes, as compared to the additional energy necessary for the path of rupture to cross a grain boundary.

To get a clearer picture of the influence of ferrite quality on machinability we must bear in mind that the constituent of steel designated by metallographists as "ferrite" is a very complex entity. It consists of alpha iron which may hold various elements in dispersion ranging from molecular (solid solution), to colloidal, and, finally, to microscopically visible particles. The range of variations in quality possible in such a complex constituent is, therefore, great. Such differences in ferrite quality manifest themselves in appreciable differences in work-hardening capacity, work-brittleness capacity, fatigue endurance, grain growth characteristics, and nature of the plastic deformation under static and dynamic loads.

It is the point of view of the writer and his associates at this time to believe that there is a rather wide range of variation in the quality of the ferrite in separate heats of steel made to the same analysis of common elements, and to further believe that this variation in ferrite quality and behavior is more important and much more fundamental than mere variation in grain-size.

It is a point not to be overlooked that, with a varying amount and varying occurrence of extraneous materials present in the base iron, there will be variations in amount and quality of material rejected to the grain boundaries. In the operation of machining, therefore, it is to be expected that there will be not only variations in the energy required to cause the path of rupture to pass through the bodies of the grains, but there will also be variations in the energy required for the path of rupture to cross the grain boundaries.

The situation is very much as would confront a hunter who walks across country, through fields and over fences. To some extent the difficulty of his progress would depend upon the number of fences which he had to climb, and, therefore, the size of the fields would be an important factor, since such size would determine the number of fences which must be climbed. Even under best conditions, however, the effort required to walk across the fields would still be appreciable as compared to that required to climb the fences. If the fields were overgrown with tangled brush, the amount of effort required to climb even a relatively high fence might be insignificant as compared to that required to cross the field.

We believe this picture to be closely parallel to the true relationship of machinability and grain-size; and we feel that until some definite evaluation of the relative energy required for the path of rupture to pass through a grain as compared to that required to cross the grain boundary, the true relationship of grain-size to machinability will remain in the field of conjecture and speculation.

DISCUSSION

Written Discussion: By N. A. Ziegler, Power Use Research Department, the West Penn Electric Company, Pittsburgh.

It is a pleasure to hear from Mr. Graham that "ferrite is a very complex entity" and that "it consists of alpha iron which may hold various elements in dispersion ranging from solid solution . . . to microscopically visible particles." It was known for some time that last traces of accidental elements present in irons and steels are not as harmless as commonly believed. From the experimental evidence presented jointly with Dr. Yensen it becomes certain that even very small variations, not exceeding a few thousandths of one per cent, in the amounts of impurities present in iron alloys, result in totally different magnetic properties.

Since this is an established fact, it would be only logical to suppose that other properties of steels, such as machinability, hardenability, etc., are equally susceptible to such small variations which usually are beyond the determination by ordinary analytical methods. It is often stated that two lots of steel of supposedly identical composition have widely different properties. Lately, as indicated by the present symposium, many of such mysteries have been solved, but there still remains a great deal of work to be done in order to obtain a better control of properties in our commercial alloys.

The solution of this problem should be approached from two directions:

1. A more refined laboratory technique, resulting in a wider knowledge and understanding of our commercial alloys.
2. A better plant technique, so that the laboratory discoveries could be successfully applied to the manufacturing and treatments of our alloys, thus resulting in a more accurate control of their properties.

Written Discussion: By H. W. McQuaid, research metallurgist, Republic Steel Corp., Massillon, Ohio.

I read with interest this paper which covers a field of the greatest practical importance. It was somewhat disappointing to find that a more definite connection was not developed between some controllable factor and machinability. It is my belief, based on a rather large scale investigation, that the grain-size of the steel as machined has a very definite bearing on tool life and finish. It is not, however, the only factor, and hence the effect of the "as-is" grain-size is often overlooked. In my opinion, the reason that a more definite effect of grain-size was not developed was because no truly coarse-grained samples were tested. In the class of steels tested, the grain-size should be fairly coarse if the

usual rolling temperatures are employed. Here, of course, we mean the grain size in the "as-rolled" condition and not the grain-size as developed by the carburizing test. The advantages of coarsening the grain by employing as high rolling temperatures as possible consistent with safe operating practice, have been well established, particularly in the field of alloy steels.

A connection between the McQuaid-Ehn grain-size and machinability is naturally only indirect, since it shows the grain-size after a prolonged heating at 1700 degrees Fahr. We are, however, interested in the grain-size after the final rolling operation, and if this is well above the coarsening range of the steel in question, as it should be, we are not concerned with the McQuaid-Ehn grain-size.

It is true of course, that the coarse-grained type as shown by the McQuaid-Ehn grain-size test can be finished at a lower rolling temperature with less adverse effect on machining than if it had been of the fine grain type as shown by the McQuaid-Ehn test. Since in this class of steels the coarsening range is approximately 1800 degrees Fahr. for the fine-grained steel, the McQuaid-Ehn test becomes of little value providing the finishing temperatures are well above 1800 degrees Fahr. The higher the finishing temperature above 1800 degrees Fahr. the better the machining results should be, regardless of the grain-size at 1700 degrees Fahr. as shown by the McQuaid-Ehn test.

Author's Reply

Mr. Ziegler's comment interested me in one way. He used a certain phraseology, and I do not mean to take exception to that phraseology because I know he used it extemporaneously. He said that certain extraneous material had been found not so harmless as had been thought. I think Mr. Ziegler and all of us probably realize that it is not a question of harmfulness, but it is a question of using these minor elements to whatever extent we can obtain control of them. The technology of not only iron and steel metallurgy, but nonferrous metals is full of cases where extremely small amounts of material have been used to great advantage. These minor elements or extraneous materials may be either harmful or helpful, and the problem of modern metallurgy consists of using them to advantage.

Mr. McQuaid's comments on high temperature rolling over other treatments are interesting. If one starts with steel in billet or bar form, one can from that point onward improve its machinability somewhat, whether by rolling temperatures, heating temperature, normalizing treatment or what not; but in certain cases improvement is obtained by going to a larger grain or to a special grain. We suspect that if it had been possible to lay aside those particular lots or samples, and go back to the original steel-making and come through again with some variation, a little more fundamental in nature, the result might have been far better than the original material in either the large or small-grained condition. In other words, to our point of view, finishing temperature, normalizing treatments, etc., are rather additive effects, superimposed upon the original quality of the material.

GRAIN-SIZE CONTROL OF OPEN-HEARTH CARBON STEELS

By S. Epstein, J. H. Nead, and T. S. Washburn

Abstract

A study of grain-size control of open-hearth steel showed that this can be assured by the use of moderate ladle additions of aluminum and by controlling the manganese and silicon content of the steel. Such control is predicated on good furnace practice, but given good furnace practice, no other recourses than those mentioned appear to be necessary.

The additions required to give a fine-grained type of steel were found not to increase the inclusion content appreciably when the aluminum was added, either in the ladle or mold. The steel became more deeply piping, however, necessitating the extra cost involved in the use of hot-topped molds.

Tensile, hardness, impact, hardenability, and grain-growth tests of the steels obtained in the study were made, and the general properties of the coarse- and fine-grained types of steel were compared and discussed. Neither type of steel has the advantage over the other for all uses; for some purposes coarse-grained would seem to be better and fine-grained for others.

Positive grain-size control in the open-hearth furnace enables the production of either type of steel at will, thus affording the steel user a dependable and uniform supply of whichever type of steel he prefers, uniformity of the supply being especially important in mass production.

Introduction

IT is now increasingly plain that the principal factor of what is known as "body" in steel is the grain-size or grain-growth characteristics. Grain-size can be easily determined and expressed numerically (1).[1] The relative clarity and definiteness thus made possible, as contrasted with the nebulous ideas implied in the pre-

[1]The figures appearing in parentheses refer to the bibliography appended to this paper.

A paper presented as part of the Grain-Size Symposium of the Sixteenth Annual Convention of the Society held in New York City the week of October 1, 1934. Of the authors, who are members of the Society, S. Epstein is metallurgist, Battelle Memorial Institute, Columbus, Ohio, J. H. Nead is chief metallurgist and T. S. Washburn is metallurgist, respectively, of the Inland Steel Co., Indiana Harbor, Indiana. Manuscript received July 14, 1934.

vailing terms, "body," "personality," "quality factors," and others, we owe largely to the pioneer work of McQuaid and Ehn (2) and Bain (3) among a host of other workers in this field. McQuaid and Ehn found that the ability of carburized steel to harden on quenching was related to the so-called "normal" or "abnormal" structure of the steel. Bain demonstrated that the grain-size was probably the controlling element in hardenability.

The abnormal structure observed by McQuaid and Ehn in the carburized layer of carburized samples differed in two respects from a normal structure. First, the grain-size was much finer; second, the hypereutectoid cementite was coalesced into relatively large masses instead of forming a fine network about the pearlite grains—even the pearlitic cementite tended to coalesce. McQuaid and Ehn rather emphasized the second feature of the abnormal structure, namely, coalescence of the cementite. In fact, it was their preoccupation with this feature of the abnormal structure which probably resulted in their use of the unfortunate term "abnormal" steel; otherwise they might have called it simply fine-grained steel. Partly on this account also they attributed "abnormality" to the presence of dissolved oxides in the steel, although they inclined to the view that "solution" of the oxides was probably of a colloidal nature. That there may be some foundation for the theory of dissolved oxides is indicated by the work of Grossmann (4), (5).

However, the feature of fine grain in the abnormal structure early came to the fore. For the production of fine grain the presence of undissolved oxides or other refractory particles that would act as nuclei for crystallization and as obstructants to grain growth during carburizing and other heat treatments seemed to supply the most acceptable mechanism. It is now generally held that submicroscopic refractory particles, mainly oxides, are the cause of fine grain.

Usually, a whole heat of steel is either coarse- or fine-grained. This fact indicates that control of grain-size resides in the furnace or deoxidation practice and is consistent with the idea that abnormality and fine grain are due to oxides. In McQuaid and Ehn's early work (2) it is mentioned that deoxidation of laboratory melts with aluminum produced a fine-grained structure. Later, Epstein and Rawdon (6) found that commercial abnormal steels were generally higher in aluminum content. They demonstrated that in open-hearth steel the addition of about one pound of aluminum per ton

in the mold made steel fine-grained, and thereupon suggested that suitable experimentation with the final deoxidation practice and aluminum additions should enable the production of either fine- or coarse-grained steel at will. Since then, announcements of the regular control of grain-size have been made by several steel manufacturers. Although the exact procedure has been surrounded by more or less secrecy, it appears rather obvious that the general means used are those already indicated. Although such statements might be interpreted to imply that rather unusual open-hearth furnace practice is involved in positive grain-size control, the work about to be described indicates that given good furnace practice, as ordinarily understood, grain-size can be regularly controlled by proper final deoxidation, augmented by moderate aluminum additions in the ladle and, if desired, in the mold.

Properties of Coarse- and Fine-Grained Steel. The first observed difference between the properties of the coarse- and fine-grained steels were those by McQuaid and Ehn on carburizing steels. They found that those of fine grain were more prone to give soft spots on quenching; in other words, they were shallower hardening or had lower hardenability. Nonuniform hardening, or the presence of soft spots, tends to increase warping. The fine-grained steels also carburized less deeply than those of coarse grain. Increased warpage and shallower and slower carburization are obvious detriments. In carburizing steel low hardenability may also be considered as a drawback, because this calls for extra care in quenching to insure a uniformly hard case. However, for many uses of medium carbon and tool steels a shallower hardening is preferred because of the greater toughness. Moreover, it soon became evident that even for carburizing steel fine grain may have decided compensatory advantages.

After carburizing, the core of a fine-grained steel is tougher than that of a coarse-grained one and shows considerably higher impact resistance (6). Both the core and the case of fine-grained steel remain so even after the prolonged heating in carburizing, so that a "core-refining" heat treatment is unnecessary and the carburized articles may be quenched directly from the carburizing temperature. For other than carburizing steel as well, the fine-grained steels have a much wider heat-treating range, because the austenite produced on heating appreciably above the critical range does not coarsen, as is the case when a coarse-grained type of steel is overheated. Fine-

grained steels are less subject to cracking on quenching than those of coarse grain and are less liable to check in grinding (8).

It has also been found that fine-grained steels are less embrittled by cold work (9). This property is of advantage in punching heavy sections (9), in heavy reductions by cold rolling (10), and its benefits are reported to have been observed in such operations as shearing and trimming (7). Fine-grained steels have higher impact resistance at low temperatures (9). It is stated (7) that the fine-grained steels flow more readily in forging than the coarse-grained steels and show less forging defects. As might be expected, the fine-grained steels have a somewhat higher elastic limit (7).

Differences in machinability between the coarse- and the fine-grained steels have been observed, and in the normalized condition the former appear to have the advantage in this respect; McQuaid (11) states that metal is removed more freely from coarse-grained steels, although those with fine grain give a smoother surface. After quenching and tempering, however, it is reported (7) that metal may be removed more rapidly from fine-grained steel, possibly because of its lower hardenability and hence lower hardness after a quench and draw at a given temperature and for a given time.

These differences in behavior between the two steels have generally been observed by steel users; the evidence for some of the alleged differences is none too certain and may leave the steel producer unconvinced. In fact, the latter, even if he has learned to control the grain-size, may, perhaps, regret that this additional burden of control is being placed upon him, on top of already too numerous specifications. However, there is no doubt that many large steel users feel certain that grain-size materially affects the behavior during fabrication and in service. For some parts coarse-grained steel may be wanted; for others, fine; in quantity production one may want, above all, uniformity of grain-size. Users, therefore, try to procure their steel from a manufacturer who can regularly control the grain-size.

The steel producer is, of course, willing to furnish the grain-size known to be best suited for the customer's requirement, where this can be done without appreciably increasing the cost. But if controlled grain-size is specified by the customer and is a matter of possible rejection, then the steel producer should be properly compensated by suitable extras for the additional cost involved.

Where grain-size is important in the customer's processing, a

definite grain-size requirement may be desirable. Otherwise difficulties in machinability, heat treatment, and fabrication may be ascribed to some other relatively unimportant factor. Or, as a result of the customer's mystification as to what is causing the trouble, he may be inclined to limit the chemical specifications too narrowly, and without benefit.

In other words, it would seem to be of advantage to both the steel producer and steel user if the vagueness implied by such terms as "quality factors" and "personality" could be largely eliminated by relating these qualities to such a readily measurable quantity as grain-size. In many instances this apparently can now be done. What grain-size is most suitable for a certain use can often only be determined by the closest co-operation between the steel maker and the steel user, and such co-operation already exists in many instances.

Control of Grain-Size

In the experiments that were made to study grain-size control, the principal variable was the amount of aluminum added in the ladle or mold. It was soon obvious that with large additions the steel would be fine-grained, and without aluminum it would be coarse-grained. However, the problem of control is not altogether so simple as this. If only to conserve the amount of aluminum used, it is desirable to add as little of this element as possible to fine-grained steels. On the other hand, aluminum is of recognized value as a final deoxidizer, and it would not always be desirable to exclude its use in coarse-grained steels. The point is that, aside from whether it is coarse- or fine-grained, the steel must be first of all good steel. What must be determined then for positive grain-size control is the minimum amount of aluminum that will always make steel fine-grained and the maximum amount of aluminum that will always still leave the steel coarse-grained.

On the assumption that the active agent in grain-size control is a suspension of tiny particles of aluminum oxide, it is obvious that the production of this chemical compound demands the presence both of metallic aluminum and of oxygen. The latter may be in the form of a reactive oxide of one or more of the metallic elements in steel, presumably iron oxide or manganese oxide, in solution in the bath so as to be uniformly disseminated and ready to produce the desired aluminum oxide in a uniform dispersion.

Thus, every factor that may affect the composition of the bath as to its content of reactive oxygen has to be considered in the control of grain-size, just as much as the amount of aluminum to be added and the conditions under which it is added.

Hence, in critical control the state of deoxidation (or more accurately, of oxidation, because reactive oxygen is needed) of the steel before the aluminum is added plays an important part. For example, it apparently takes less aluminum to make a higher carbon steel fine-grained than a lower carbon steel. Similarly, but much more decidedly, the silicon content of the steel strongly influences the amount of aluminum required to make the steel fine-grained, this being appreciably less in steel of higher silicon content. The manganese content plays a rôle similar to silicon, though its effect is not so pronounced.

How the state of deoxidation of the steel before the aluminum addition affects the outcome may, perhaps, be pictured as follows: If the steel is highly oxidized, the aluminum added is more likely to form large inclusions instead of sub-microscopic oxide particles, because of the greater amount of oxygen present and because the fluxing action of the excess iron oxide would tend to increase the size of the inclusions. Presumably, the larger inclusions would not make for fine grain as effectively as the sub-microscopic particles. After preliminary deoxidation, however, usually with manganese, or silicon, the particles that form on adding aluminum would be small enough to cause fine grain. In the case of previous deoxidation with silicon, moreover, the aluminum might reduce some of the silicates present to form alumina; under these circumstances the particles formed would also tend to be very small. For this and other reasons it may be stated here that it is preferable to add the aluminum after, rather than before, the other deoxidizers.

From the foregoing it is evident that the melting procedure or general furnace practice in the open-hearth must have its influence on the final grain-size. Mention has already been made of the announced (7) recourses (aside from the usual slag analyses and measurements of the rate of carbon drop) for checking the furnace practice to insure positive grain-size control. The development of such special tests may be of equal interest in general open-hearth practice. However, it can be concluded as a result of this study that a good, well-established furnace practice need not be tampered with for grain-size control. Positive control can be accomplished merely

by fitting final aluminum additions to the condition of the bath and the carbon, manganese, and silicon content of the steel.

Experimental Results

The effect of aluminum on grain-size is illustrated by Figs. 1, 2, and 3 of carburized samples of the first four heats (series I) of Table I. For the McQuaid-Ehn tests the samples were carburized at 1725 degrees Fahr. (940 degrees Cent.) for ten hours. The open-hearth heats referred to were 125 and 90 tons, and the ingots weighed approximately five to eight tons. The aluminum was added to the mold as small chunks, distributed uniformly through the ingot from bottom to top as it was being poured.

Fig. 1 shows that the addition of 1 pound of aluminum per ton in the mold made the steel fine-grained both in heat 16864, to which no aluminum had been added in the ladle, and in heat 27109, to which 1 pound per ton had been added. Although it tended to reduce the grain-size somewhat, 1 pound of aluminum per ton in the ladle in heat 27109 did not produce a fine grain. It is obvious from this that mold additions of aluminum are more effective than ladle additions in causing fine grain.

For the experiment illustrated in Fig. 3, it had been intended to choose a heat of similar carbon and manganese content to heat 16874, but with 0.20 per cent instead of 0.13 per cent silicon. In such a heat as with the higher silicon, the addition of 1 pound of aluminum per ton in the ladle would have made the heat fine-grained; with the lower silicon, however, a further addition of 4 ounces per ton of aluminum in the mold was required. Likewise, in series II of Table I, in heats with higher carbon and manganese contents and with silicon ranging from 0.11 to 0.17 per cent, the addition of 1 pound per ton of aluminum in the ladle was enough to make the heats fine-grained. It will be observed, however, that in series IIA_2, heat 21781, which had the lowest carbon, manganese, and silicon content in this series, the full effect of fine grain was not accomplished by 1 pound per ton of aluminum in the ladle; the case was fine, but the core was coarse. It was concluded as a result of these tests that to insure fine grain in steels ranging from 0.15 to 0.50 per cent carbon, with about 0.50 per cent manganese, by means of about 1 pound per ton ladle addition of aluminum, the steel should contain not much less than 0.15 per cent silicon. With an appreciably higher manga-

Table I

Results of Experimental Series of Open-Hearth Heats for Study of Grain-Size Control

Series	Group	Heat	Ladle Analysis Per Cent C	Mn	Si	Furnace Additions	Aluminum in Ladle Oz. per Ton	Ingot No.	Aluminum in Mold Oz. per Ton	McQuaid-Ehn Grain Size	Depth of Carburized Layer, In.
I	A	16864	0.18	0.40	0.03	SiMn	none	2	none	Coarse	
								3	16	Fine	
	B	27109	0.12	0.35	0.03	SiMn	16	2	none	Coarse	
								3	16	Fine	
	C	16874	0.21	0.47	0.20	SiMn	none	2	none	Coarse	
								3	8	Mixed	
								4	12	Fine case, coarse core	
								5	16	Fine	
	D	4970	0.18	0.42	0.13	SiMn	16	2	none	Coarse	0.045
								3	4	Fine	0.043
								4	8	Fine	0.043
II	A1	15125	0.44	0.76	0.14	SiMn	16	2	none	Fine	0.053
								3	4	Fine	0.053
								4	8	Fine	0.053
	A2	21781	0.32	0.58	0.11	SiMn	16	2	none	Fine case, coarse core	0.061
								3	4	Fine	0.055
								4	8	Fine	0.053
	B1	23949	0.48	0.68	0.16	Spiegel +FeSi	16	2	none	Fine	0.061
								3	4	Fine	0.073
								4	8	Fine	0.072
	B2	23950	0.44	0.64	0.15	Spiegel +FeSi	16	2	none	Fine	0.076
								3	4	Fine	0.076
								4	8	Fine	0.076
	C	5621	0.37	0.68	0.17	SiMn	6	2	none	Coarse to medium	
								3	4	Fine	
								4	8	Fine	
								5	12	Fine	
III	A	2176	0.18	0.57	0.03	SiMn	20	2	none	Coarse	0.061
								3	4	Fine	0.043
								4	8	Fine	0.051
								5	12	Fine	0.045
	B	4887	0.18	0.45	0.03	SiMn	6	2	none	Coarse	0.039
								3	4	Coarse	0.041
								4	8	Coarse	0.043
								5	12	Fine	0.037
	C	18473	0.19	0.44	0.10	SiMn	12	2	none	Coarse	0.060
								3	4	Coarse	0.055
								4	8	Mixed	0.055
								5	12	Fine	0.043
	D	4889	0.15	0.40	0.06	SiMn	6	2	none	Coarse	0.053
								3	4	Coarse	0.060
								4	8	Coarse	0.057
								5	12	Fine	0.055
IV	A1	22523	0.41	0.58	0.14	Spiegel +FeSi	16 as Al-Si-Fe Alloy	2	none	Fine	0.055
								3	4	Fine	0.057
								4	8	Fine	0.055
	A2	26785	0.43	0.57	0.19	Spiegel +FeSi	8 as Al-Si-Fe Alloy	2	none	Coarse	0.073
								3	4	Fine	0.054
								4	8	Fine	0.058

nese content the silicon may be lower and under favorable furnace conditions may be as low as 0.07 per cent.

Of course, if mold additions of aluminum are countenanced, the silicon content could be somewhat lower. With about 0.12 per

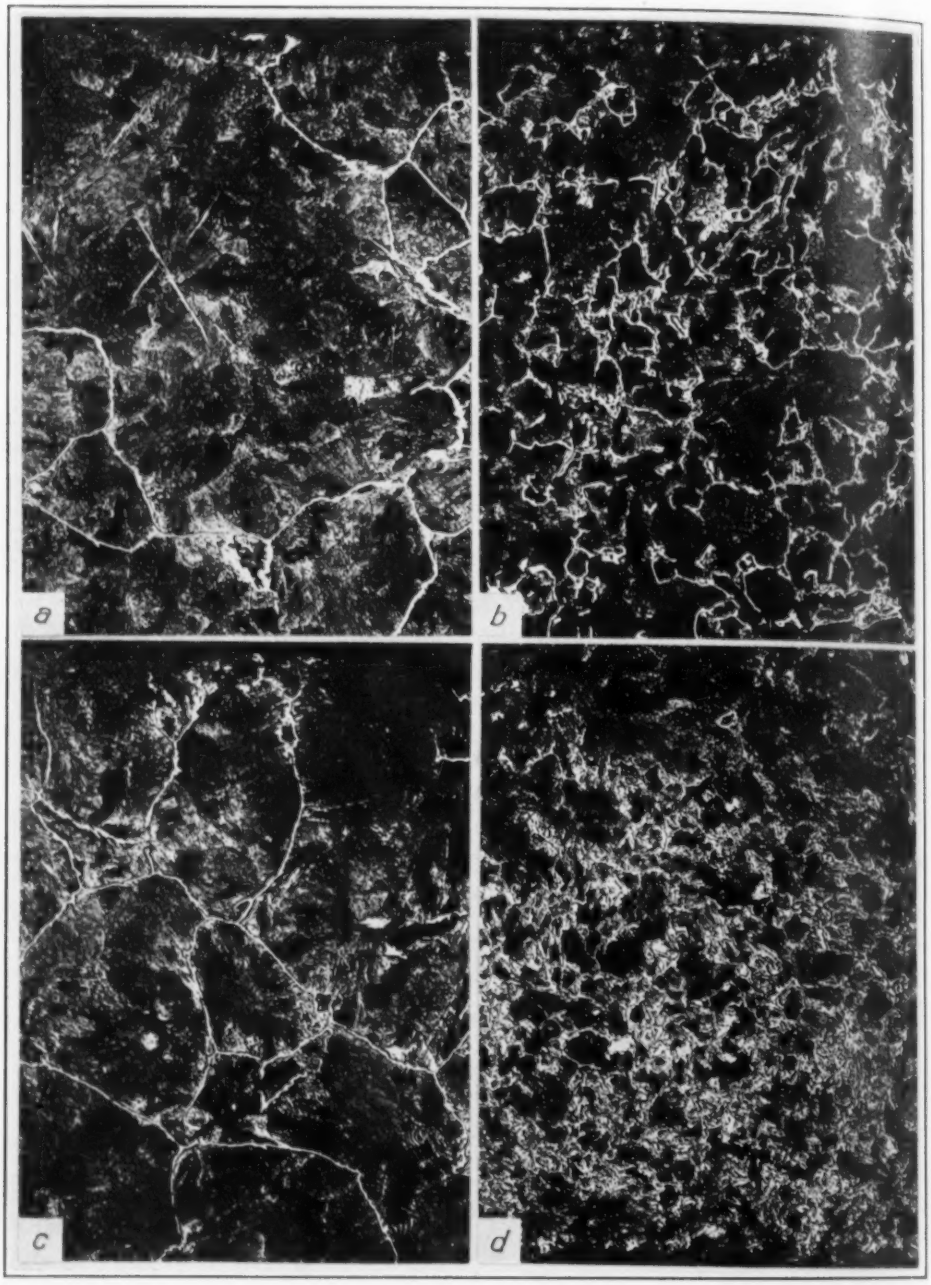


Fig. 1—McQuaid-Ehn Tests Showing Effect of Aluminum in Ladle and Mold. × 100.

	Treatment and Result	
	a and c	b and d
Heat 16864 0.18 Per Cent Carbon, 0.40 Per Cent Manganese, 0.03 Per Cent Silicon No Aluminum in ladle	No Aluminum in mold	1 Pound Aluminum per ton in mold
Heat 27109 0.12 Per Cent Carbon, 0.35 Per Cent Manganese, 0.03 Per Cent Silicon 16 ounces per ton Aluminum in Ladle	Coarse Grain	Fine Grain

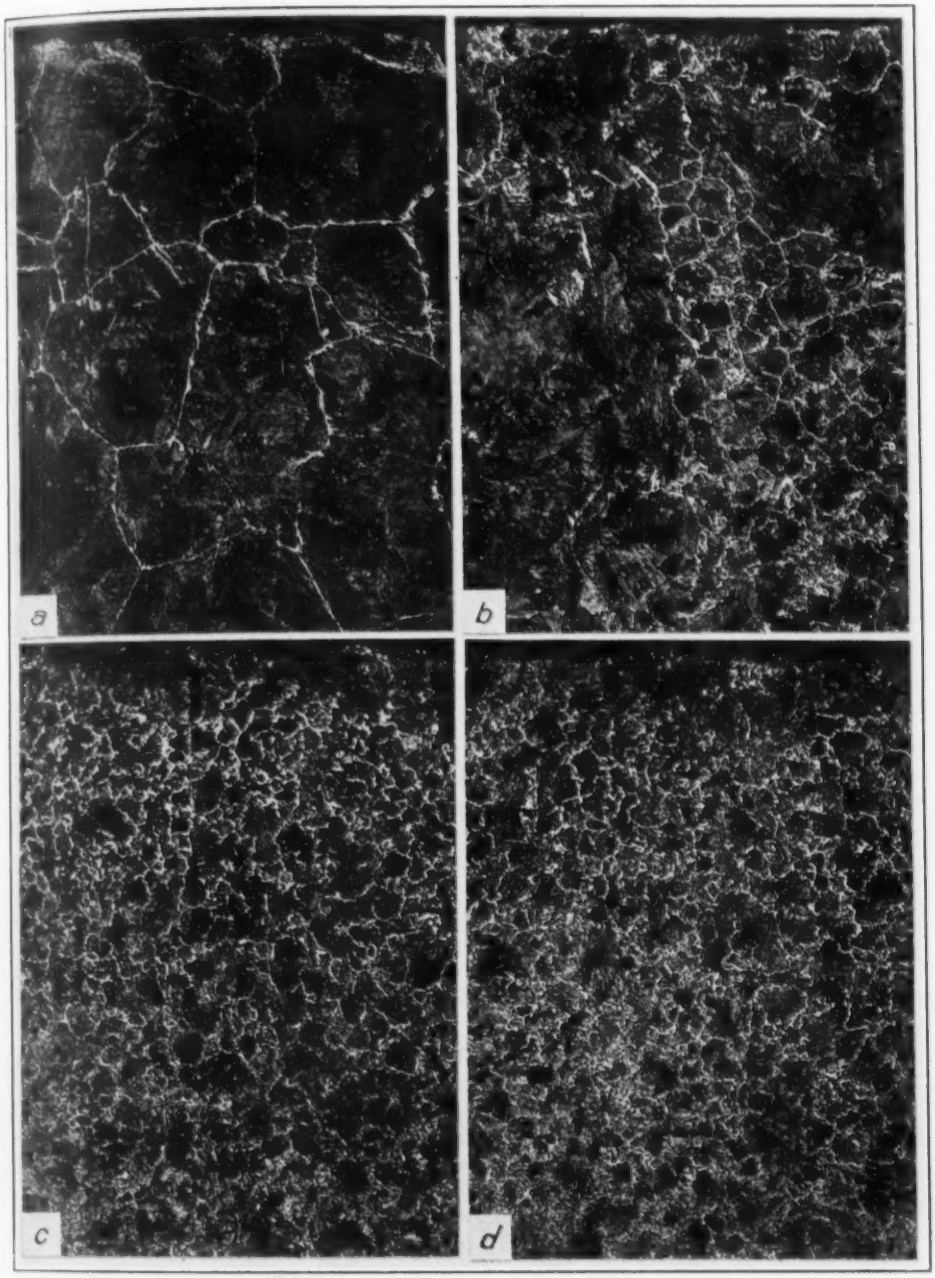


Fig. 2—McQuaid-Ehn Tests Showing Effect of Aluminum in Ladle and Mold. × 100.

Heat 16874: 0.21 Per Cent Carbon, 0.47 Per Cent Manganese, 0.20 Per Cent Silicon; no Aluminum in Ladle.

a—No Aluminum in Mold, Coarse Grain.
b—8 Ounces Per Ton Aluminum in Mold, Mixed Grain.
c—12 Ounces Per Ton Aluminum in Mold, Fine Case, Coarse Core.
d—16 Ounces Per Ton Aluminum in Mold, Fine Grain.

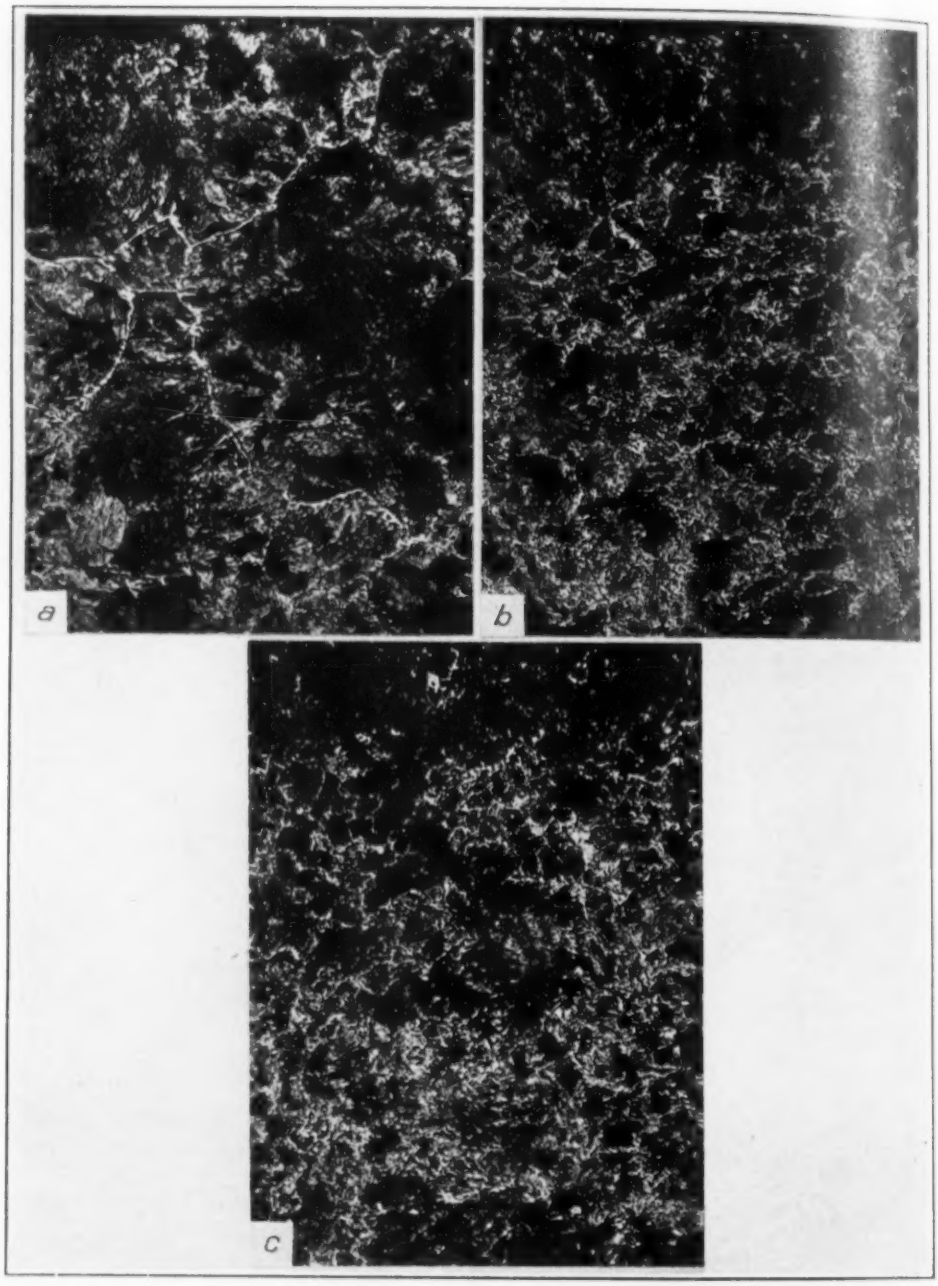


Fig. 3—McQuaid-Ehn Tests Showing Effect of Aluminum in Ladle and Mold. × 100. Heat 4970: 0.18 Per Cent Carbon, 0.42 Per Cent Manganese, 0.13 Per Cent Silicon; 16 Ounces Per Ton Aluminum in Ladle.

a—No Aluminum in Mold, Coarse Grain.
b—4 Ounces Per Ton Aluminum in Mold, Fine Grain.
c—8 Ounces Per Ton Aluminum in Mold, Fine Grain.

cent silicon, 1 pound per ton of aluminum could be added in the ladle, after which an addition of 4 ounces per ton in the mold should insure fine grain. The experiments made did not indicate such mold additions of aluminum to be harmful. Where only a part of the heat is ordered fine-grained and a coarse grain is desired for the remainder, it would be a simple matter, therefore, to make aluminum additions in the mold for the fine-grained portion of the heat. In general, however, it is not regarded as desirable to use aluminum in the mold, and the aim, therefore, in the grain-size control experiments was to produce fine grain by ladle additions only.

The limits below which the aluminum additions must be kept to obtain a coarse-grained steel are also fairly clear. In general, less than ½ pound per ton in the ladle to a steel of 0.10 per cent silicon content or less should produce a steel of coarse grain.

The relations between carbon, manganese, and silicon content, ladle additions of aluminum, and the resultant grain-size in a considerable number of open-hearth heats are shown in Table II. It will be observed that there are virtually no exceptions to the rules already indicated for insuring grain-size control.

Although the silicon content plays such an important part in producing fine grain by means of ladle additions of aluminum, it should be emphasized that a fine grain cannot be obtained by even a high-silicon content in the absence of aluminum. This is illustrated in Fig. 4 of the carburized layer of a structural silicon steel which contained 0.45 per cent silicon, but which still had a moderately coarse grain. Several similar high-silicon structural steels are shown in Table II, all with relatively coarse grain.

Intermediate Grain-Size. An important limitation of the method of grain-size control by means of aluminum alone is that in general a grain of intermediate size is difficult to obtain. This is illustrated by Fig. 2, b and c. Insufficient aluminum to quite give a fine grain does not result in an intermediate size, but rather in a mixed grain, as in Fig. 2b, consisting of coarse and fine grains, or in a fine-grained case and a coarse-grained core, as in Fig. 2c. It is not readily possible to obtain an intermediate grain regularly like No. 4 on the A. S. T. M. grain-size chart, in plain carbon steels. Large additions of manganese and silicon tend to reduce the grain-size (compare Fig. 4 with Fig. 1a), so that by this means, and possibly with small aluminum additions, an intermediate grain should regularly be obtainable. However, the silicon and manganese contents would

Table II
Grain-Size of Heats Made in Regular Production

Carbon Range, %	Heat	Grade of Steel	Ladle Analysis Per Cent C	Mn	Si	Furnace Additions	Aluminum Additions, Oz. per Ton Ladle	Mold	Grain-Size Large or Small 2-3	Medium Mixed 5-7
0.15–	13918	Commercial	0.26	0.60	0.10	Spiegel + FeSi	..	...	*	
0.29	21627	Commercial	0.28	0.62	0.11	Spiegel + FeSi	..	...	*	
	26636	Commercial	0.28	0.65	0.13	Spiegel + FeSi	..	...	*	
	4839	Forging	0.25	0.71	0.19	Spiegel + FeSi	..	...	*	
	9198	Forging	0.27	0.83	0.17	Spiegel + FeSi	..	...	*	
	17090	Forging	0.15	0.92	0.12	SiMn	..	...	*	
	16874	Forging	0.21	0.48	0.20	SiMn	..	...	*	
	4889	Commercial	0.15	0.40	0.06	SiMn	6	0.6	*	
	18473	Forging	0.19	0.44	0.10	SiMn	12	...	*	
	27089	Forging	0.22	0.49	0.14	Spiegel + FeSi	..	0.6		*
	10268	Forging	0.19	0.39	0.10	Spiegel + FeSi	..	2.6		*
	19261	Forging	0.22	0.47	0.12	Spiegel + FeSi	16	...		*
	20018	Forging	0.22	0.83	0.20	Spiegel + FeSi	16	...		*
0.30–										
0.49	10280	Forging	0.32	0.52	0.09	Spiegel + FeSi	..	1/3	*	
	3207	Forging	0.35	0.60	0.12	Spiegel + FeSi	..	1/6	*	
	9223	Forging	0.44	0.66	0.18	Spiegel + FeSi	..	1/6	*	
	4818	Forging	0.47	0.60	0.19	Spiegel + FeSi	..	1/3	*	
	2101	Forging	0.33	0.60	0.16	Spiegel + FeSi	11	...	*	
	24880	Forging	0.37	0.71	0.07	Spiegel + FeSi	16	...	*	
	26672	Forging	0.43	0.55	0.07	Spiegel + FeSi	16	...	*	
	26571	Forging	0.38	0.55	0.13	Spiegel + FeSi	16	...	*	
	24103	High-silicon structural	0.39	1.28	0.50	Spiegel + FeSi	..	...	*	
	25025	High-silicon structural	0.37	1.12	0.42	Spiegel + FeSi	..	...	*	
	20072	High-silicon structural	0.35	1.02	0.43	Spiegel + FeSi	..	...	*	
	22333	High-silicon structural	0.38	1.12	0.46	Spiegel + FeSi	..	...	*	
	3201	High-silicon structural	0.37	1.35	0.44	Spiegel + FeSi	..	...		*
	24881	Forging	0.40	0.90	0.07	Spiegel + FeSi	16	...		*
	21502	Forging	0.41	0.90	0.09	Spiegel + FeSi	16	...		*
	22170	Forging	0.40	0.70	0.09	Spiegel + FeSi	16	...		*
	24941	Forging	0.36	0.74	0.10	Spiegel + FeSi	16	...		*
0.30–	26458	Forging	0.34	0.83	0.15	Spiegel + FeSi	16	...		*
0.49	24862	Forging	0.38	0.80	0.18	Spiegel + FeSi	16	...		*
	22180	Forging	0.39	0.89	0.18	Spiegel + FeSi	16	...		*
	25093	Forging	0.35	0.79	0.19	Spiegel + FeSi	16	...		*
	25085	Forging	0.33	0.78	0.19	Spiegel + FeSi	16	...		*
	23785	Forging	0.38	0.67	0.19	Spiegel + FeSi	16	...		*
0.50–	20080	Commercial	0.58	0.74	0.02	Spiegel + FeSi	12	1/4	*	
0.69	19218	Commercial	0.61	0.75	0.06	Spiegel + FeSi	12	1/4	*	
	4869	Forging	0.55	0.55	0.16	Spiegel + FeSi	12	1/3	*	
	20981	Commercial	0.58	0.78	0.07	Spiegel + FeSi	12	1/4		*
	10278	Forging	0.52	0.55	0.15	Spiegel + FeSi	12	1/3		*
	26717	High-silicon structural	0.57	1.43	0.50	Spiegel + FeSi	..	...		*
	22398	Silico-Manganese spring	0.59	0.90	1.82	Spiegel + FeSi	..	...		*
	2191	Commercial†	0.53	0.73	0.16	Spiegel + FeSi	16	...		*
	5590	Commercial	0.59	0.69	0.18	Spiegel + FeSi	16	...		*
	9253	Commercial	0.53	0.87	0.20	Spiegel + FeSi	16	...		*
	10313	Commercial	0.54	0.72	0.21	Spiegel + FeSi	16	...		*
	4893	Commercial	0.58	0.80	0.24	Spiegel + FeSi	16	...		*
0.70–	23861	Spring	0.79	0.75	0.16	.....	..	...	*	
0.89	26640	Spring	0.72	0.70	0.17	.....	..	...	*	
	19343	Spring	0.72	0.82	0.18	.....	..	...	*	
	20102	Spring	0.79	0.74	0.20	.....	..	...	*	
	14447	Spring	0.78	0.68	0.15	.....	16	...		*
	22476	Spring	0.74	0.68	0.16	.....	16	...		*
0.90–	16885	Spring	0.95	0.42	0.26	.....	..	...	*	
1.10	25048	Spring	0.96	0.38	0.27	.....	..	...	*	
	26831	Spring	0.99	0.27	0.30	.....	..	...	*	
	13920	Spring	0.92	0.33	0.31	.....	..	...	*	

†An unusually low yield was obtained on these heats because the aluminum addition required to give fine grain increased the depth of piping.

probably have to be so high as to remove the steel from the category of ordinary plain carbon steel. The problem of the regular production of a steel of intermediate grain-size calls for further study. This does not mean that the coarse-grained, plain-carbon steels have the extreme grain-size corresponding to No. 1 on the A. S. T. M. chart, or that the fine-grained steels are extremely fine, corresponding to

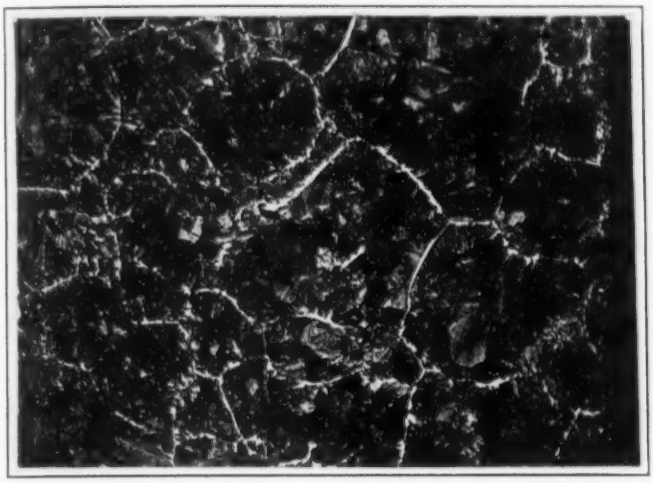


Fig. 4—Moderately Coarse Grain in Carburized Layer of Structural Silicon Steel. × 100.
Heat 22333: 0.38 Per Cent Carbon, 1.12 Per Cent Manganese, 0.45 Per Cent Silicon.

No. 8. The coarser-grained steels observed in this study usually ranged from No. 2 to No. 3 and the fine-grained steels from No. 5 to No. 7. The aluminum additions never gave a grain-size finer than No. 7.

Piping in Fine-Grained Steel. Because of the amounts of silicon and aluminum required, fine-grained steel tends to be more deeply piping than the ordinary commercial grade of steel. The latter is not usually hot-topped, so that a tendency toward deeper piping would reduce the yield. The purpose, therefore, of making the experimental heats Series III in Table I was to determine whether a fine-grained steel could be made that would not tend to pipe deeply in an ingot without a hot top. One of the lines of attack was to add less aluminum in the ladle and more in the mold, because the latter is more effective than the former in causing fine grain. Less silicon, with more aluminum in the ladle, was also tried, as indicated in Series III of Table I.

To obtain quantitative results of the depth of piping would have necessitated successive croppings from the top blooms of the ingots

until the end of the pipe was reached. This was not done. However, deep-etch tests definitely indicated that the minimum silicon and aluminum contents required to cause fine grain made the steel more deeply piping than ordinary commercial steel.

This result is of considerable commercial importance. Frequently a steel user will order an ordinary commercial grade (non-hot-topped) steel even if he wishes to use it for forging purposes. This, of course, is his privilege and his own responsibility. However, if he goes a step further and orders *fine-grained* commercial steel, he will be burdening the steel maker with the increased expense involved not only in an additional item of control and in the increased amounts of silicon and aluminum required, but also in the necessarily lower yield obtained, if the product is to be cropped until no piping is in evidence. Although the tests made were not exhaustive, it would appear that because of the depth of piping it would not be economical to make fine-grained steels except in hot-topped molds. Thus, a fine-grained steel should really belong not with the commercial grade (non-hot-topped) steels, but with the higher grade forging steels which bear the price extra for forging quality, and which are customarily made in hot-topped molds.

Depth of Case. The depth of case obtained for a given carburizing treatment is a fairly important consideration in carburizing. As already stated, it is well known that the case tends to be less deep in fine-grained steels than in those of coarse grain. This was confirmed by microscopic measurements of the depth of case (to the extremity of the eutectoid zone) of the experimental series of heats and ingots shown in Table I. In coarse-grained steels the case averaged 10 to 20 per cent deeper than in the fine-grained steels. It will be noted that once the mold addition of aluminum had made the steel fine-grained, any further addition did not tend to decrease the depth of case. The case depth was naturally deeper in the higher carbon steels. Manganese may also increase the case depth. Possibly, therefore, the comparative shallowness of the case in the fine-grained steel may be overcome by raising the manganese content, and, perhaps, even by going to a slightly higher carbon content; with the greater toughness of the fine-grained steel a higher carbon content in the core should not be objectionable.

Microscopic Inclusion Counts and Oxide Analyses

The use of aluminum as a deoxidizer is generally considered

rather objectionable for cleanness of steel. As the method of obtaining fine grain herein described depends on the use of aluminum, it was thought desirable to determine the cleanness of the experimental heats made, and of the ingots of these heats to which varying amounts of aluminum had been added in the mold.

Cleanness of the steel was determined by the microscopic method described by Epstein (12) and by the electrolytic extraction method using magnesium iodide as the electrolyte, described by Scott (13).

In the microscopic method the inclusions in the rolled steel are determined in a 2-inch round section rolled or forged from the billet. All of the larger inclusions in a longitudinal section of the 2-inch round bar ½ inch square in area are counted at 100 magnifications. The section is taken through the axis of the round and extends ½ inch along the length of the bar from the center to the outside. Fig. 5 shows the section examined. Only the larger inclusions are counted. These are weighted according to size, and hence to their probable adverse influence on the mechanical properties, as indicated in the reference (12) cited. The inclusions can readily be divided into sulphides, silicates, and oxides by their appearance. The total weights for all of the larger inclusions in the sample give the index number of cleanliness for that sample.

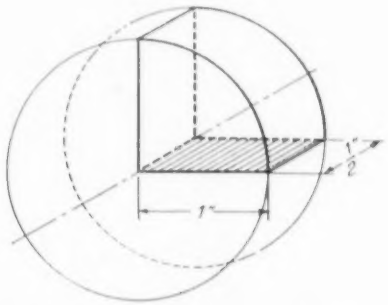


Fig. 5 — Sampling the 2-Inch Round for Microscopic Examination for Inclusion Count.
All of the Larger Inclusions in the Cross-Hatched Section are Counted at 100 Magnification.

The microscopic inclusion counts and the oxide analyses of the experimental heats are listed in Table III. It will be observed that in general the correspondence between the inclusion counts and the oxide analyses is fairly good. Thus, both methods decidedly indicate more oxides in the bottom than in the middle of the ingot.

The four higher carbon heats (15125, 21781, 23949, and 23950)

Table III

Microscopic Inclusion Counts and Oxide Analyses of Experimental Series of Open-Hearth Heats for Study of Grain-Size Control

Heat Number and Description	Ingot No.	Al in Mold Oz. per Ton	McQuaid-Ehn Grain-Size	Position in Ingot	Microscopic Inclusion Counts[a]				Oxide Analyses[b]			
					Sulphides	Silicates	Oxides	Total Oxides	SiO_2	Al_2O_3	MnO	Total Oxides
4970, Group ID, 0.18% C, 0.42% Mn, 0.13% Si	2	none	Coarse	Top	0	0	6	6				
				Middle	1	0	8	8				
				Bottom	0	0	21	21				
SiMn in furnace	3	4	Fine	Top	2	1	6	7				
				Middle	0	0	10	10				
				Bottom	0	0	26	26				
Al in ladle 16 oz. per ton	4	8	Fine	Top	0	0	30	30				
				Middle	5	0	5	5				
				Bottom	0	0	30	30				
2176, Group IIIA, 0.18% C, 0.57% Mn, 0.03% Si	2	none	Coarse	Top	8	0	2	2				
				Middle	5	0	2	2	1.9	7.0	0.1	9
				Bottom	1	0	48	48	3.2	13.9	0.1	17.2
SiMn in furnace	3	4	Fine	Middle	5	0	9	9				
				Bottom	0	0	41	41				
Al in ladle 20 oz. per ton	4	8	Fine	Middle	7	0	2	2				
				Bottom	0	0	30	30				
	5	12	Fine	Middle	2	0	3	3	2.1	9.5	0.1	11.7
				Bottom	0	0	20	20	1.3	11.0	0.1	12.4
4887, Group IIIB, 0.18% C, 0.45% Mn, 0.03% Si	2	none	Coarse	Top	12	10	0	10				
				Middle	0	40	0	40	19.7	5.0	24.3	49
				Bottom	1	31	0	31	20.5	4.8	21.9	47.2
SiMn in furnace	3	4	Coarse	Top	10	1	14	15				
				Middle	4	4	21	25				
				Bottom	0	3	0	3				
Al in ladle 6 oz. per ton	4	8	Coarse	Top	9	0	12	12				
				Middle	1	0	14	14				
				Bottom	0	0	29	29				
	5	12	Fine	Top	0	0	51	51				
				Middle	1	0	15	15	1.8	16.8	0.1	18.7
				Bottom	0	0	13	13	2.6	23.5	0.3	26.4
4889, Group IIID, 0.15% C, 0.40% Mn, 0.06% Si	2	none	Coarse	Middle	8	0	5	5	8.8	15.7	7.0	31.5
				Bottom	1	6	19	25	11.8	24.2	5.9	41.9
SiMn in furnace	3	4	Coarse	Middle	3	0	7	7				
				Bottom	0	0	48	48				
Al in ladle 6 oz. per ton	4	8	Coarse	Middle	4	0	11	11				
				Bottom	0	0	92	92				
	5	12	Fine	Middle	5	0	13	13	1.5	9.1	0.1	10.7
				Bottom	0	0	38	38	1.2	13.7	0.0	14.9
15125, Group IIA_1, 0.44% C, 0.76% Mn, 0.14% Si	2	none	Fine	Middle	6	0	0	0	3.1	5.2	.1	8.4
				Bottom	0	0	10	10	.9	6.7	.1	7.7
SiMn in furnace	3	4	Fine	Middle	3	0	1	1				
				Bottom	4	0	4	4				
Al in ladle 16 oz. per ton	4	8	Fine	Middle	5	0	5	5	1.5	3.1	.1	4.7
				Bottom	2	0	8	8	1.2	6.3	.1	7.6
21781, Group IIA_2, 0.32% C, 0.58% Mn, 0.11% Si	2	none	Fine case, coarse core	Top	1	0	18	18				
				Middle	1	0	0	0	1.0	9.4	.1	10.5
				Bottom	2	0	18	18	.7	7.8	.1	8.6
SiMn in furnace	3	4	Fine	Top	3	0	24	24				
				Middle	5	0	12	12				
				Bottom	0	0	5	5				
Al in ladle 16 oz. per ton	4	8	Fine	Top	5	0	8	8				
				Middle	1	0	1	1	1.1	7.8	.3	9.2
				Bottom	0	0	2	2	1.1	7.3	.3	8.7

Table III (Continued)

Microscopic Inclusion Counts and Oxide Analyses of Experimental Series of Open-Hearth Heats for Study of Grain-Size Control

Heat Number and Description	Ingot No.	Al in Mold Oz. per Ton	McQuaid-Ehn Grain-Size	Position in Ingot	Microscopic Inclusion Counts[a]				Oxide Analyses[b]			
					Sulphides	Silicates	Oxides	Total Oxides	SiO_2	Al_2O_3	MnO	Total Oxides
23949, Group IIB_1, 0.48% C, 0.68% Mn, 0.16% Si Spiegel + FeSi in furnace Al in ladle 16 oz. per ton	2	none	Fine	Top	10	0	9	9	2.3	7.7	1.0	11
				Middle	2	0	5	5	0.2	9.4	0.1	9.7
				Bottom	4	0	19	19	3.4	10.9	0.1	14.4
	3	4	Fine	Top	39	0	1	1				
				Middle	6	1	0	1				
				Bottom	7	0	10	10				
	4	8	Fine	Middle	7	0	10	10	4.1	10.9	1.2	16.2
				Bottom	19	0	8	8	0.4	7.3	...	7.7
23950, Group IIB_2, 0.44% C, 0.64% Mn, 0.15% Si Spiegel + FeSi in furnace Al in ladle 16 oz. per ton	2	none	Fine	Top	26	0	0	0	3.3	11.4	1.1	15.8
				Middle	2	0	2	2	2.1	8.4	0.1	10.6
				Bottom	0	0	17	17	0.4	11.8	...	12.2
	3	4	Fine	Top	30	0	0	0				
				Middle	10	0	13	13				
				Bottom	4	0	11	11				
	4	8	Fine	Top	11	0	2	2	2.2	5.6	0.1	7.9
				Middle	5	0	7	7	1.2	6.0	0.1	7.3
				Bottom	8	0	13	13	3.1	8.7	0.2	12.0
5621, Group IIC, 0.37% C, 0.68% Mn, 0.17% Si SiMn in furnace Al in ladle 6 oz. per ton	2	none	Coarse to Medium	Top	17	0	7	7				
				Middle	8	0	22	22	1.5	7.9	.7	10.1
				Bottom	5	0	34	34	1.7	9.0	.7	11.4
	3	4	Fine	Middle	19	0	9	9				
				Bottom	3	0	32	32				
	4	8	Fine	Top	2	4	22	26				
				Middle	5	0	9	9				
				Bottom	0	0	41	41				
	5	12	Fine	Top	0	0	24	24				
				Middle	4	0	19	19	1.5	11.5	.1	13.1
				Bottom	6	0	45	45	1.3	16.5	.1	17.9
22523, Group IVA_1, 0.41% C, 0.58% Mn, 0.14% Si Spiegel + FeSi in furnace Al in ladle 16 oz. per ton as Al-Si-Fe alloy	2	none	Fine	Top	0	0	2	2				
				Middle	16	2	3	3	2.4	12.9	.2	15.5
				Bottom	0	0	67	67	1.2	7.4	.1	8.7
	3	4	Fine	Top	4	0	1	1				
				Middle	5	0	6	6				
				Bottom	1	0	37	37				
	4	8	Fine	Top	0	0	26	26				
				Middle	0	0	2	2	3.9	4.3	.5	8.7
				Bottom	0	0	80	80	3.1	9.1	.1	12.3
26785, Group IVA_2 0.43% C, 0.57% Mn, 0.19% Si Spiegel + FeSi in furnace Al in ladle 8 oz. per ton as Al-Si-Fe alloy	2	none	Coarse	Top	6	9	10	19				
				Middle	5	0	3	3	1.9	3.4	.7	6
				Bottom	0	0	33	33	2.1	9.7	.7	12.5
	3	4	Fine	Top	0	5	16	21				
				Middle	1	0	5	5				
				Bottom	0	0	52	52				
	4	8	Fine	Top	0	0	10	10				
				Middle	0	0	7	7	1.3	10.6	.7	12.6
				Bottom	0	0	44	44	1.1	15.0	.6	16.7

[a]These figures are index number given to total weight of the inclusions observed in the section of the sample examined.

[b]Figures given in thousandths of a per cent.

were treated with 1 pound of aluminum per ton in the ladle. These heats, fine-grained without any further mold additions, were the cleanest heats in the experimental series. The further aluminum additions of 4 and 8 ounces per ton in the mold seemed to have no noticeable effect in increasing the number of inclusions.

Heat 5621, similar in carbon content to the four heats discussed, but with only 6 ounces per ton of aluminum in the ladle, had an appreciably higher inclusion content. Here again the further additions of aluminum, up to 12 ounces per ton in the mold, had no noticeable effect in increasing the number of larger inclusions. In fact, it may be stated that in none of the heats examined did mold additions of aluminum up to 12 ounces per ton have any appreciable effect in increasing the number of large inclusions.

The four lower carbon heats (4970, 2176, 4887, 4889) in Table III had an appreciably higher inclusion content than the four clean higher carbon heats referred to. In these lower carbon heats the ladle additions of aluminum varied from 6 to 20 ounces per ton. That the relatively higher inclusion content of these heats was not caused by the ladle additions of aluminum is indicated by the fact that heats 4970 and 2176 with ladle additions of 16 and 20 ounces per ton, respectively, did not differ greatly in inclusion content from heats 4887 and 4889; in fact, if anything, the two heats with the higher ladle additions were somewhat cleaner. The cause for the higher inclusion content of these heats probably lies in the more highly oxidized condition of the bath associated with their lower carbon content.

It should be noted that the ladle addition of aluminum required to give fine grain, approximately 1 pound per ton in a steel of the proper carbon, manganese, and silicon content, is really very moderate. Frequently in regular open-hearth practice as much as 1 pound per ton of aluminum in the ladle is added as an ordinary deoxidizer and scavenger, and with no thought of grain-size control.

Heats 22523 and 26785, both of higher carbon, had a distinctly higher inclusion content than the clean* higher carbon heats. The ladle additions of aluminum in the case of these two heats were the same or less than in the clean heats, but the aluminum was added in the form of an aluminum-silicon-iron alloy, instead of as stick

*The term "clean" is meant here purely in a relative sense and is used only for comparison purposes. As a matter of fact, all of the experimental heats examined were relatively clean and fully acceptable as commercially clean steel of their respective grades.

aluminum. The relatively high inclusion counts of these two heats may have been due to the use of this alloy. It was probably not due to the use of spiegel plus ferrosilicon, instead of silicomanganese in the furnace, as this practice produced a very clean steel in the case of heat 23950.

The results of the inclusion counts on heat 4887 are of interest. In this heat 6 ounces of aluminum per ton was added in the ladle. With no mold addition of aluminum, ingot No. 2 showed a considerable number of silicates and no oxides. Ingot No. 3 with 4 ounces of aluminum per ton in the mold, in addition to that in the ladle, showed less silicates and some oxides. Ingot No. 4 with 8 ounces of aluminum per ton in the mold showed no silicates, only oxides. Evidently the mold additions of aluminum reduced the silica to form alumina. The accompanying oxide analyses in Table III indicate that the inclusions recognized under the microscope as silicates or oxides are not pure SiO_2 or Al_2O_3; the silicates may contain an appreciable amount of Al_2O_3 and the oxides some SiO_2.

To summarize, the moderate ladle additions of aluminum required to cause fine grain had no tendency to increase the number of detrimental inclusions. Even the mold additions of aluminum used, up to 12 ounces per ton, showed no such tendency. The use of an aluminum-silicon-iron alloy instead of stick aluminum as a means of making ladle additions of aluminum appeared to offer no advantages. Indeed, the two aluminum-silicon-iron alloy-treated heats were not as clean as comparable heats treated with the equivalent amount of aluminum in the form of stick aluminum. But enough data are not available on this to draw any definite conclusions.

Ingot Surface. Of course, the inclusion count is not the only measure of steel quality. Often surface, which means the presence or absence of surface seams and other defects, is at least of equal importance. The factor of surface could not be so readily determined as the inclusion content. The few data available, however, such as chipping costs, do not indicate that either the ladle or mold additions of aluminum required to give fine grain appreciably harmed the surface condition of the steel.

Deoxidizers Other Than Aluminum. It should be mentioned here that besides aluminum other strong deoxidizers like titanium, zirconium, and vanadium are also known to cause fine grain. The first two give rise to tiny oxide particles, much as aluminum does, and there is some reason to believe that vanadium also exerts its

grain-refining effect through the mechanism of producing discrete, sub-microscopic particles rather than purely through its action in lowering the amount of dissolved iron or manganese oxide. However, somewhat greater additions of these elements appear to be required than for aluminum, and they are considerably more expensive. The question is whether titanium and zirconium, for example, give sufficiently better results, if any, than aluminum, to justify their greater cost, particularly for plain carbon steel. As already indicated, the amount of strong deoxidizer required to give fine grain is not much greater than that frequently used in killed steel. The problem of what deoxidizer to use for grain-size control, therefore, does not differ much from the problem of what strong deoxidizer to use in ordinary steel-making, and here, of course, aluminum is much more widely employed than titanium and zirconium, for example.

Further tests are now in progress to determine whether zirconium or titanium in conjunction with aluminum are more satisfactory for grain-size control than aluminum alone.

Tests of Coarse- and Fine-Grained Steels

The material of the experimental series of heats given in Table I, with the varying additions of aluminum in the molds, proved especially advantageous for comparing the properties of coarse- and fine-grained steel, because in several of the heats coarse- and fine-grained steel was available in adjacent ingots of the same heat. This gave assurance that extraneous factors, such as undue variations in chemical composition, were absent and that grain-size alone was the main variable in the comparisons made.

The tests made were tensile, hardness, impact, hardenability, and heating to determine at what temperature coarse-grain growth ensued in the fine-grained types of steel. It would have been of interest also to compare the fatigue resistance, machinability, and forgeability of the coarse- and fine-grained steels. Such tests are contemplated but could not be undertaken in the time available for the preparation of this report.

Tensile, Hardness, and Impact Tests. Most of the experimental heats were rolled into heavy billets, so that tests in the as-rolled condition could not be conveniently made. This was, therefore, done only on the two heats 18473 and 4889. Aside from these tests in the as-rolled condition, billets from the middle and bottom of coarse- and

fine-grained ingots of all of the heats (except heat 18473 which was rolled directly into ⅜-inch thick channels) were forged down to 1-inch rounds and tested after heat treatment.

It is well known (14), and this was confirmed by observations on several of the experimental heats and on other heats examined in this study, that usually there is practically no difference in grain-size between the coarse- and fine-grained types of steel in the as-rolled condition. The difference in grain-size only becomes apparent after a carburizing or similar higher temperature heat treatment. Hence, to bring out the inherent differences in grain-size between the coarse- and fine-grained steels, the 1-inch round specimens to be tested were heat treated (the coarse- and fine-grained samples exactly alike), one set being air-cooled after heating 1 hour at 1625 degrees Fahr. (885 degrees Cent.), and another set being air-cooled after heating 2 hours at 1750 degrees Fahr. (995 degrees Cent.). The specimens were then machined to size. The tensile specimens were 0.505 inch diameter, 2-inch gage length. The impact specimens were standard Charpy bars with the drilled key-hole notch and were machined from the middle of the air-cooled 1-inch round bars; the cold working of 9 per cent was done with a die, as described by Bauer (15).

To list the tensile, hardness, and impact test results completely would take too much space. Only the averaged results are, therefore, shown in Tables IV and V. Table IV gives the results of the heats in which both coarse- and fine-grained ingots were available, and Table V gives those heats that were fine-grained after the ladle addition of aluminum, but in which the ingots that had a further mold addition of 8 ounces per ton were also tested.

It will be observed in Table IV that the yield strength of the finer-grained steel was appreciably higher than that of the coarser-grained steel, this being especially noticeable after the 1750 degrees Fahr. (955 degrees Cent.) treatment which brought out a decided difference in grain-size between the two types of steel. After this treatment, in the steels ranging from 0.15 to 0.19 per cent carbon the yield strength was about 5000 pounds per square inch higher in the fine- than in the coarse-grained steel. After the 1625 degrees Fahr. (885 degrees Cent.) treatment, however, when the difference in grain-size between the coarse- and fine-grained steels was less, there was a less difference in yield strength. In the as-rolled condition there was practically no difference. The tensile strength,

Table IV
Averaged Tensile, Hardness, and Impact Tests of Coarse- and Fine-Grained Steels

Carbon Range, %	Heat Treatment	Position in Ingot	Tensile Tests: Yield Strength Lbs. Per Sq. In. Coarse-Grain Type	Yield Strength Lbs. Per Sq. In. Fine-Grain Type	Tensile Strength Lbs. Per Sq. In. Coarse-Grain Type	Tensile Strength Lbs. Per Sq. In. Fine-Grain Type	Elongation in 2 Inches, % Coarse-Grain Type	Elongation in 2 Inches, % Fine-Grain Type	Reduction of Area, % Coarse-Grain Type	Reduction of Area, % Fine-Grain Type	Hardness Tests Rockwell B Coarse-Grain Type	Hardness Tests Rockwell B Fine-Grain Type	Impact Tests Energy Absorbed, Ft.-Lbs. Coarse-Grain Type Not Worked	Coarse-Grain Type Cold Worked 9%	Fine-Grain Type Not Worked	Fine-Grain Type Cold Worked 9%
0.15-0.19 (5 heats)	Air-cooled 1625° F.	Middle	39,400	42,000	61,200	61,000	35.4	36	64.6	66.1	61.9	64.0	43.4	37.3	47.6	43.8
		Bottom	37,300	41,500	60,300	60,500	35.9	36.4	66.1	67.3	64.1	63.6	45.0	39	47.5	44
	Air-cooled 1750° F.	Middle	32,600	39,100	60,600	60,300	32	35.5	62.2	66.9	63.2	63.2	40.5	23.5	45.5	38.1
		Bottom	33,200	38,600	59,500	59,500	34	33.8	62.4	64.9	63.7	63.2	41	27.7	46.2	40.3
0.32-0.43 (3 heats)	Air-cooled 1625° F.	Middle	49,100	51,100	86,700	82,000	28.3	31	56.9	58.1	85.2	83.6	24.7	20.2	28	23.6
		Bottom	44,200	46,800	82,800	78,700	30	30	53.4	58.3	83.8	80.9	26.2	20	31	26.7
	Air-cooled 1750° F.	Middle	45,500	47,400	90,000	85,300	25.3	29	49.4	54	86.2	82.7	21.8	13.5	27.6	25.2
		Bottom	43,700	47,800	86,100	81,300	26.3	29.3	51.7	56.1	82.2	81.2	27.6	20.3	31.3	27.0

Table V
Averaged Tensile, Hardness, and Impact Tests of Fine-Grained Steels with No Aluminum in the Mold and with 8 Ounces per Ton of Aluminum in the Mold

Carbon Range, %	Heat Treatment	Position in Ingot	Tensile Tests: Yield Strength Lbs. Per Sq. In. Ingot 2[a]	Yield Strength Lbs. Per Sq. In. Ingot 4[b]	Tensile Strength Pounds Per Sq In. Ingot 2[a]	Tensile Strength Pounds Per Sq In. Ingot 4[b]	Elongation in 2 Inches, % Ingot 2[a]	Elongation in 2 Inches, % Ingot 4[b]	Reduction of Area, % Ingot 2[a]	Reduction of Area, % Ingot 4[b]	Hardness Tests (Rockwell B) Ingot 2[a]	Hardness Tests (Rockwell B) Ingot 4[b]	Impact Tests Engery Absorbed, Ft.-Lbs. Ingot 2[a] Not Worked	Ingot 2[a] Cold Worked 9%	Ingot 4[b] Not Worked	Ingot 4[b] Cold Worked 9%
0.41-0.48 (4 heats)	Air-cooled 1625° F.	Middle	53,300	53,800	94,200	92,700	26.5	27	51.2	51.9	87.3	88.1	21.2	16.2	20.8	19
		Bottom	52,100	52,300	88,900	88,300	28.2	29.2	53.4	55.3	84.5	85.3	24.1	21	24.1	22.1
	Air-cooled 1750° F.	Middle	53,900	53,200	96,600	94,800	23.2	26.2	48.1	49.4	87.2	87.2	20.9	14.8	22.9	20.4
		Bottom	51,000	51,600	90,500	88,000	26.2	27.8	50.9	52.4	86.1	84.1	22.6	18.0	27.4	23.9

[a]Ingot 2, no Al in mold.
[b]Ingot 4, 8 oz. per ton Al in mold.

gots
ples
segr
and
treat
steels
grain
area
tests
formi
a sor
excep
howe

however, did not differ between the coarse- and fine-grain steels, except in the higher carbon range, where the coarse-grain steel showed a somewhat higher tensile strength. The hardness in general conformed to the tensile strength. The elongation and reduction of area were slightly higher in the fine-grain steels. Although impact tests did not show the large differences between the coarse- and fine-grain steels expected, they were definitely higher in the finer-grain steels, particularly after the 1750 degrees Fahr. (955 degrees Cent.) treatment and after cold working 9 per cent.

It will be observed that the samples from the bottom of the ingots were somewhat softer, more ductile, and tougher than the samples from the middle of the ingot. This is no doubt a result of carbon segregation in the ingot. The uniform difference between the middle and bottom of the ingots served to confirm the accuracy of the tests.

This is also confirmed by Table V comparing the test data on fine-grained steels with no aluminum in the mold and with 8 ounces of aluminum per ton in the mold. The difference in properties between them was negligible, indicating that once a fine grain is obtained, further additions of aluminum have little effect. It should be noted, however, that the impact and ductility values were slightly higher in the samples with the mold additions of aluminum. This may possibly have been caused by a slight diminution in grain size, although such differences were not recognized under the microscope. In the McQuaid-Ehn tests the fine-grained ingots without aluminum in the mold, and with aluminum in the mold, had about the same grain size, 6 to 7 according to the A. S. T. M. grain-size chart.

The following conclusions may be drawn. Normalized fine-grained steels have an appreciably higher yield strength and impact resistance and somewhat higher ductility than coarse-grained steels. The tensile strength is about the same in both, unless in higher carbon samples the rate of cooling is fast enough (as it may be on air cooling samples 1-inch or less in diameter) for the greater lag in Ar_1 (6) of the coarse- over the fine-grained steel to come into play. In that case, the coarse-grained steel may have a somewhat higher tensile strength. If the rate of cooling is fast enough, the increased hardenability of the coarse-grained steel may presumably also result in its showing a higher yield strength than the fine-grained steel.

In general, these differences in properties are in favor of the fine-grained steels. However, they are predicated on an actual difference in grain-size between the two, such as exists after carburiz-

ing and higher temperature normalizing treatments. They do not exist directly after rolling when there is generally no difference in grain-size between the coarse- and the fine-grained types of steel. On the other hand, for the important class of steel which is heat treated above the critical range as in normalizing, annealing, and quenching (aside from carburizing and higher temperature normalizing which have already been mentioned), there is always more likelihood of obtaining a somewhat coarser grain in a coarse-grained type of steel than in a fine-grained type. For such heat treated steel, therefore, the advantageous properties of the fine-grained type of steel appear to deserve consideration.

Hardenability Tests. As it has been repeatedly demonstrated (6), (3), (7) that the fine-grained steels have lower hardenability than those with coarse grain, only a few hardenability tests were made of the material of the experimental heats. Coarse- and fine-grained specimens of one of the lower and one of the higher carbon heats were chosen. These were tested after quenching in water and in brine (10 per cent solution), in the as-rolled condition, and after carburizing. The samples were in the form of 1¾-inch rounds, 3 inches long. They were heated for hardening in an electric muffle furnace in an air atmosphere.

Fig. 6 shows the difference in thickness of the hardened layer between the coarse- and fine-grained samples of heat 26875, containing 0.43 per cent carbon and 0.57 per cent manganese. After brine-quenching from 1500 degrees Fahr. (815 degrees Cent.), the hardened layer was considerably deeper in the coarse-grained steel, but the hardened layer was uniformly hard in both types of steel. After water-quenching, the hardened layer in the fine-grained steel was not only shallower but also non-uniform, containing soft spots.

For most quenched parts the attainment of uniform surface hardness without soft spots is the main consideration. In the other hardenability tests, therefore, the surface hardness of the 1¾-inch diameter cylinders mentioned was measured—six equally spaced longitudinal rows of Rockwell readings being taken around the circumference of the cylinder, five equally spaced readings in each row. In the quenched lower carbon steel in the as-rolled condition the hardness was very low, and the Rockwell B readings taken were merely averaged. In the other samples the results were expressed as an "index number of soft spots" (6). That means, only the readings below 60 Rockwell C were considered as being an indication

Table VI

Hardenability Tests of Coarse- and Fine-Grained Steels

Heat Number and Description	AS-FORGED SAMPLES — Average Rockwell B Hardness — Brine-Quenched from 1625° F. — Coarse-Grained Type	AS-FORGED SAMPLES — Average Rockwell B Hardness — Brine-Quenched from 1625° F. — Fine-Grained Type	CARBURIZED SAMPLES — Index Number of Soft Spots* Quenched from 1450° F. into Water — Coarse-Grained Type	CARBURIZED SAMPLES — Index Number of Soft Spots* Quenched from 1450° F. into Water — Fine-Grained Type	CARBURIZED SAMPLES — Index Number of Soft Spots* Quenched from 1450° F. into Brine — Coarse-Grained Type	CARBURIZED SAMPLES — Index Number of Soft Spots* Quenched from 1450° F. into Brine — Fine-Grained Type
4887, 0.18% C, 0.45% Mn, 0.03% Si						
Al in ladle, 6 oz. per ton						
Al in mold:						
Ingot 2, none, coarse	101	92.5	7	10	0	0
Ingot 5, 12 oz. per ton, fine	99	90.4	7	14	0	0

Heat Number and Description	Index Number of Soft Spots* Quenched from 1500° F. into Water — Coarse-Grained Type	Index Number of Soft Spots* Quenched from 1500° F. into Water — Fine-Grained Type	Index Number of Soft Spots* Quenched from 1500° F. into Brine — Coarse-Grained Type	Index Number of Soft Spots* Quenched from 1500° F. into Brine — Fine-Grained Type
26785, 0.43% C, 0.57% Mn, 0.20% Si				
Al (as Al-Si-Fe alloy in ladle) 8 oz. per ton				
Al in mold:				
Ingot, 2 none, coarse	17	44	5	23
Ingot 4, 8 oz. per ton, fine	..	..	0	20

Heat Number and Description	CARBURIZED SAMPLES — Index Number of Soft Spots* Quenched from 1450° F. into Water — Coarse-Grained Type	CARBURIZED SAMPLES — Index Number of Soft Spots* Quenched from 1450° F. into Water — Fine-Grained Type	CARBURIZED SAMPLES — Index Number of Soft Spots* Quenched from 1450° F. into Brine — Coarse-Grained Type	CARBURIZED SAMPLES — Index Number of Soft Spots* Quenched from 1450° F. into Brine — Fine-Grained Type
26785, 0.43% C, 0.57% Mn, 0.20% Si				
Al (as Al-Si-Fe alloy in ladle) 8 oz. per ton				
Al in mold:				
Ingot, 2 none, coarse	1	3	0	0
Ingot 4, 8 oz. per ton, fine	2	14	0	0

*The index number of soft spots was obtained by totaling weights of the hardness readings under 60 Rockwell C for the 30 readings on the surface of the 1¼-inch diameter specimens, as follows: reading 55 to 60, weight 1; 50 to 55, 2; 45 to 50, 3; under 45, 4.

of soft spots, and these readings were weighted as follows: 55 to 60, weight 1; 50 to 55, weight 2; 45 to 50, weight 3; below 45, weight 4. The total weights for all of the 30 readings then composed the index number of soft spots.

Table VI gives the results. It will be observed that in the lower carbon steel in the as-rolled condition, brine-quenched from 1625 degrees Fahr. (885 degrees Cent.), the fine-grained steel averaged lower in hardness than the coarse-grained steel. In the higher carbon steel, in the as-rolled condition, the fine-grained steel had a higher index number of soft spots, which means more soft spots, than the coarse-grained steel both after water- and brine-quenching from 1500 degrees Fahr. (815 degrees Cent.). There were fewer soft spots after brine- than after water-quenching.

In the carburized samples of both the lower and higher carbon steel the fine-grained samples showed more soft spots on water-quenching from 1450 degrees Fahr. (790 degrees Cent.). However,

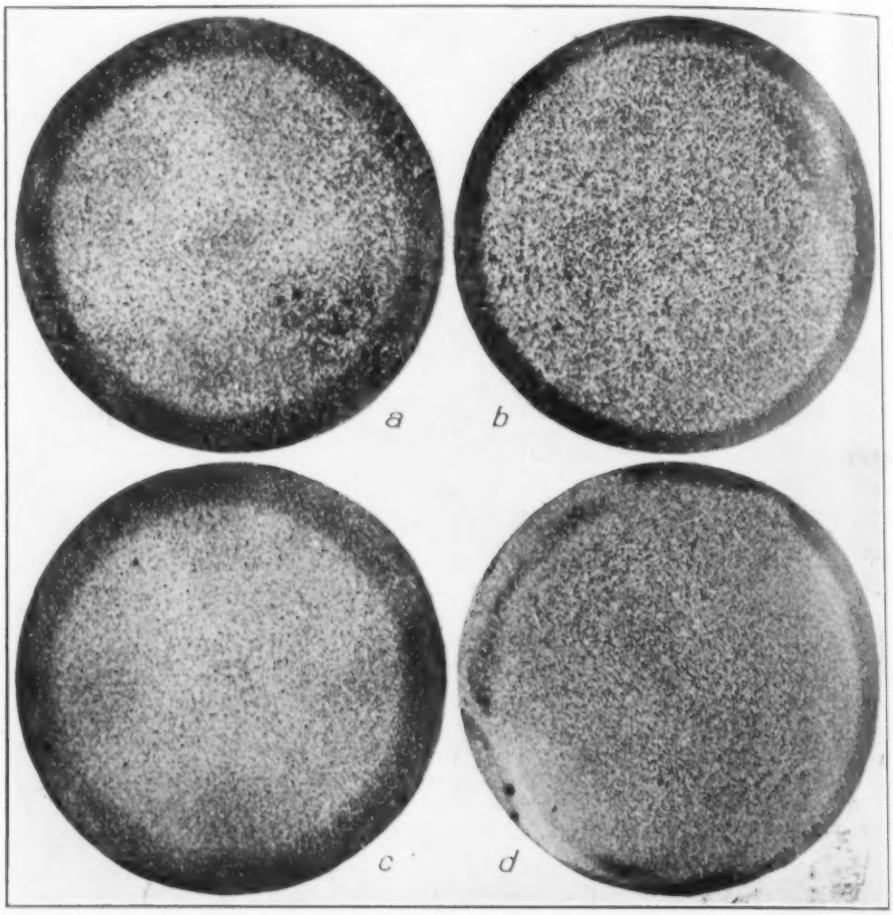


Fig. 6—Difference in Hardenability Between Coarse (a and c) and Fine (b and d) Grain Steel.

a and b were Brine-Quenched;
c and d were Water-Quenched.

Cylinders 1¾ Inches in Diameter from Heat 26785 (0.43 Per Cent Carbon, 0.57 Per Cent Manganese, 0.20 Per Cent Silicon) Quenched in Brine from 1500 Degrees Fahr., Sectioned and Deeply Etched in Hot 1 : 1 HCl; the Darker Rings Show the Depth of Hardening. In the Darker Rings the Rockwell C Hardness Ranged from 45 to 60; in the Lighter (Softer) Areas the Hardness Averaged About 30.

After Brine-Quenching the Hardened Layer was Uniform; After Water-Quenching Large Soft Spots (Lighter Areas) are Shown in the Outer Ring of the Fine-Grained Steel.

after brine-quenching from 1450 degrees Fahr. no soft spots appeared on either the fine- or coarse-grained samples.

The conclusion may be drawn that the fine-grained steels have a definitely lower hardenability and greater tendency to form soft spots than the coarse-grained steel. This is apparent on ordinary water-quenching. However, if due precautions for uniform and drastic quenching are observed, such as heating in a salt bath containing cyanide and quenching in a water-pressure spray or brine or sodium

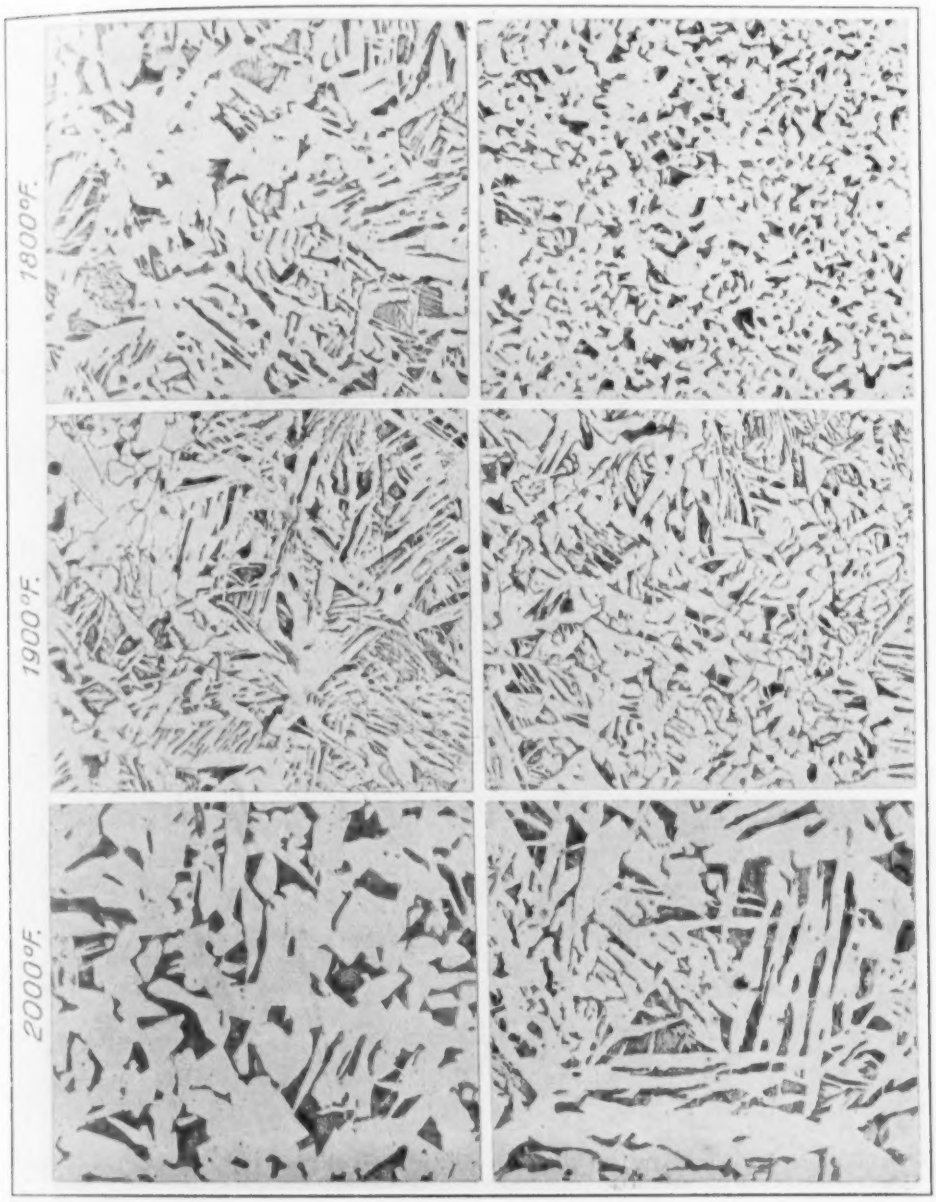


Fig. 7—Grain Growth on Heating a Coarse- and a Fine-Grained Type of Steel. Left Group Coarse-Grained Type. Right Group Fine-Grained Type. × 100.
On Heating to 1800 Degrees Fahr., the Coarse-Grained Steel Became Definitely Coarse. The Fine-Grained Steel Remained Fine at 1800 Degrees Fahr., Showed Evidences of Coarsening at 1900 Degrees Fahr., and Became Definitely Coarse at 2000 Degrees Fahr.

hydroxide solution, uniform hardening and freedom from soft spots can be obtained in both coarse- and fine-grained steel.

Manganese is known to increase hardenability. It would be of interest, therefore, to determine exactly what the manganese con-

tent of a fine-grained steel should be for it to have equal hardenability with a coarse-grained steel. Apparently, 0.25 per cent more manganese in the fine-grained steel is not enough to give it equal hardenability with the coarse-grained steel (7), and probably appreciably greater amounts than this would be required.

Temperature of Coarse-Grain Growth in Fine-Grained Steel

It is of interest to know at what heating temperature a fine-grained type of steel begins to show evidence of grain growth. Thus, for many higher temperature heat treatments the finer-grained steels have a wider heat treatment range in which coarse-grain growth does not occur, but still there is, of course, an upper limit to this range. On the other hand, McQuaid (11) has brought out that one of the purposes of normalizing is to coarsen the grain somewhat, for easier machining. For this purpose it would be desirable that grain coarsening should be initiated at not too high a temperature. Again, it has been pointed out (14) that if grain growth could be restrained to beyond the temperatures used in rolling, it might be possible to obtain a fine-grained steel as rolled, which at present cannot be done. To accomplish this the grain-size of the fine-grained steel would have to remain fine on heating to as high or higher than 2200 degrees Fahr. (1205 degrees Cent.). Grossmann (14) has reported results on fine-grained specimens which remained fine on heating almost to this temperature, namely, to 2150 degrees Fahr. (1175 degrees Cent.).

A series of heat treatments was, therefore, made of a coarse- and of a fine-grained sample from the same heat (16874), containing 0.21 per cent carbon, 0.47 per cent manganese, and 0.20 per cent silicon. The coarse-grained sample was not treated with aluminum in the mold, but the fine-grained sample got 12 ounces per ton of aluminum in the mold. The samples were heated for 2 hours at 1800, 1900, and 2000 degrees Fahr. (980, 1040, 1095 degrees Cent.).

As may be seen in Fig. 7 the samples of the coarse-grained type of steel became fairly coarse on heating to 1800 degrees Fahr. (980 degrees Cent.). The sample of the fine-grained type of steel remained rather fine at this temperature, although signs of coarsening had begun to appear. After heating to 1900 degrees Fahr. (1040 degrees Cent.), coarsening of the grain was more in evidence, and

on heating to 2000 degrees Fahr. (1095 degrees Cent.), the sample was definitely coarse.

The grain-growth characteristics on heating different types of fine-grained steel have only been touched upon here. Much work remains to explore the interesting possibilities involved.

Summary

1. A study of grain-size control of open-hearth steel showed that given ordinarily good furnace practice positive production of coarse- or of fine-grained carbon steel at will can be attained by a controlled ladle addition of aluminum and a controlled reactive oxygen content, which in turn depends largely on the carbon, manganese, and silicon content of the steel. For a range of carbon steel between 0.15 and 0.50 per cent, and for a manganese content of 0.50 per cent and above, and a silicon content of about 0.15 per cent and above, a fine-grained type of steel was invariably obtained, with the furnace practice used, by means of an aluminum addition in the ladle of about 1 pound per ton. With the furnace practice employed, a coarse-grained steel is obtained by omitting the aluminum, or else using considerably less aluminum; in general, an addition of less than about ½ pound per ton of aluminum in the ladle to a steel of less than about 0.10 per cent silicon content results in a coarse-grained type of steel.

2. Production of intermediate grain-size to schedule in a nominally "plain carbon" steel, with standard basic open-hearth melting practice, requires reduction of reactive oxygen below the amount that is utilized in making the fine-grained steel. This requires a higher content of manganese and/or silicon and is really equivalent to making a low-alloy steel.

3. Microscopic inclusion counts and oxide analyses indicate that the moderate amounts of aluminum required to give a fine-grained type of steel did not increase the number of larger inclusions, or adversely affect the cleanness of the steel. This was true when the aluminum was added either in the ladle or mold.

This would indicate that satisfactory grain-size control by means of aluminum can be accomplished without detriment to the quality of the steel. However, further experiments are being conducted on the use of titanium or zirconium in conjunction with aluminum, and

possibly some such combination may be more satisfactory than aluminum, for some purposes.

4. The amount of aluminum and silicon required to give fine grain results in a fully killed deeply piping steel. Such steel for best results must be poured in hot-topped molds and should bear some price extra for this type of practice.

5. Tests of the mechanical properties of the coarse- and fine-grained steels obtained indicated that after a heating which produced a difference in grain size between the coarse- and fine-grained steels, followed by air-cooling, the fine-grained steels had a higher yield strength, slightly higher ductility, and higher impact resistance than the coarse-grained steels, and had about equal tensile strength.

6. The hardenability of the fine-grained steels was distinctly lower than that of the coarse-grained steels. After carburizing, however, uniform hardening of the fine-grained steels could be obtained by quenching in brine.

Acknowledgments

We wish to acknowledge our indebtedness to the Inland Steel Co., who sponsored this work and in whose plant the experiments were conducted, for permission to publish the results. Acknowledgment is also made to F. W. Scott, chemist, and R. E. Parkinson, metallurgist, of the Inland Steel Co., and to H. B. Kinnear of the Battelle Memorial Institute, for the oxide analyses and valued aid in the metallographic work.

References

1. Tentative Standard Grain-Size Chart. American Society for Testing Materials Standards, 1933, Pt. I, Metals, p. 898.
2. W. H. McQuaid and E. W. Ehn, "Effect of Quality of Steel on Case Carburizing Results," *Transactions*, American Institute of Mining and Metallurgical Engineers, Vol. 67, 1922, p. 341.
3. E. C. Bain, "Factors Affecting the Inherent Hardenability of Steel," Transactions, American Society for Steel Treating, Vol. 20, 1932, p. 385-428.
4. M. A. Grossmann, "Oxygen Dissolved in Steel, and Its Influence on the Structure," Transactions, American Society for Steel Treating, Vol. 16, 1929, p. 1-19.
5. M. A. Grossmann, "Oxygen in Steel," Transactions, American Society for Steel Treating, Vol. 18, 1930, p. 601-631.
6. S. Epstein and H. S. Rawdon, "Steel for Case-Hardening Normal and Abnormal Steel," Bureau of Standards, *Journal* of Research, Vol. 1, 1928, p. 423-466.
7. Carnegie Steel Co., Carnegie-Controlled Carbon Steels, 1933.

8. E. S. Davenport, E. L. Roff, and E. C. Bain, "Microscopic Cracks in Hardened Steel, Their Effects and Elimination," TRANSACTIONS, American Society for Metals, Vol. 22, 1934, p. 289.

9. S. Epstein, "A Study of the Embrittlement of Hot-Galvanized Structural Steel," *Proceedings*, American Society for Testing Materials, Vol. 32, Part 2, 1932, p. 293-379.

10. G. R. Brophy and L. L. Wyman, "Factors Influencing the Annealing of Cold Rolled Strip Steel," TRANSACTIONS, American Society for Steel Treating, Vol. 21, June, 1933, p. 532-556.

11. H. W. McQuaid, "Trend in Requirements for Plain and Alloy Steels," *Metal Progress*, Jan., 1934, p. 15-18.

12. S. Epstein, "A Suggested Method of Determining the Cleanness of a Heat of Steel," *Metals and Alloys*, Vol. 2, 1931, p. 186-191.

13. F. W. Scott, "Extraction of Slag and Oxide Inclusions in Iron or Steel," *Industrial and Engineering Chemistry*, Analytical Ed., Vol. 4, 1932, p. 121-125.

14. M. A. Grossmann, "On Grain Size and Grain Growth," TRANSACTIONS, American Society for Steel Treating, Vol. 21, 1933, p. 1079-1104. Discussion by S. Epstein.

15. O. Bauer, "Steel of Slight Susceptibility to Aging," *Zeitschrift, Bayerischer Revisions-Verein*, Vol. 32, 1928, p. 23. See also reference 9.

DISCUSSION

Written Discussion: By W. H. Woodhall, metallurgist, Harrison Steel Castings Co., Attica, Indiana.

The authors of this paper are to be commended very highly on the work presented. Bringing up the subject of adding aluminum to steel takes us back several years in our experience when we used to add aluminum to all our heats. We make plain carbon steel by the acid open-hearth process. In our experience, aluminum has helped us but very little in obtaining steel of good quality. We make this assertion because most of our castings must undergo heat treatment. These castings are of the grade S.A.E. 1045. Such erratic results of hardness were accomplished after a water quench, that we finally left out all aluminum from our heats, and have never gone back to that practice. In fact, our customers using this grade of steel objected very strongly to practice of adding aluminum.

We believe that good quality steel can be made without aluminum additions, especially where the steel must be heat treated afterward.

Written Discusion: By G. F. Comstock, metallurgical engineer, Titanium Alloy Manufacturing Co., Niagara Falls, N. Y.

This paper gives a large amount of interesting data which will repay careful study, and the authors have performed a useful service in publishing it. Some of their conclusions, however, as expressed in paragraph 3 of the Summary, do not seem to be strictly in accordance with the results reported.

The statement that aluminum additions in the mold did not adversely affect the cleanness of the steel is based apparently on the data of Table III, and is given considerable emphasis both at the top of page 960, and my repetition in the middle paragraph of page 961. A careful study of the microscopic inclusion counts fails, however, to justify this conclusion.

To bring out clearly and concisely the effect of aluminum additions in the mold, the sum of the "total oxide" count with no mold addition was compared, for each heat separately, with the similar sum for the maximum mold addition of aluminum. In heats 2176 and 23949 the top sample without aluminum was omitted since no top sample with aluminum is reported in the paper. Then the percentage increase in the inclusion count for the ingot with aluminum in the mold is readily computed.

	Total Oxide Count		
Heat No.	No Mold Addition	Maximum Aluminum	Per Cent Increase With Maximum Aluminum
4970	35	65	86
2176	50	23	—54
4887	81	79	—2
4889	30	51	70
15125	10	13	30
21781	36	11	—69
23949	24	18	—25
23950	19	22	16
5621	63	88	40
22523	72	108	50
26785	55	61	11

In this way it is readily seen that seven heats out of eleven were made dirtier by the aluminum mold addition, the sum of the positive percentage increases in dirtiness being 303 as against 150 negative, or an average increase in oxide inclusion count for all heats of 14 per cent. This certainly shows that aluminum in the mold made this still dirtier in general, although there were some individual exceptions as frequently happens in work of this nature. A similar computation based on the 8 ounce per ton mold addition of aluminum instead of the maximum for each heat, gave practically the same results.

Several factors which must have influenced these results were in favor of the aluminum-treated steels. One was the decision to count only the larger inclusions. It is well known that alumina inclusions are small, so that it must be assumed that many of them were not counted, although streaks of fine inclusions are of course just as apt to cause cracks and similar defects as single elongated slag fibers. If alumina inclusions were not counted just because they were small, the statements regarding the cleanness of these steels should be more carefully qualified.

Another factor influencing these results in favor of the aluminum-treated ingots is the absence of data regarding the top portions of many of the ingots. Out of seven heats for which the cleanness of the top part of the aluminum-treated ingots is reported, there are only two (21781 and 26785) in which the top sample of the ingot without aluminum was dirtier than the middle sample, while in six of the seven heats the top sample of the ingot with maximum aluminum addition was dirtier than the middle sample of the same ingot. This shows that when aluminum was used in the mold the tops of the ingots tended to be dirtier than their middle portions, and therefore the omission of top samples of the aluminum-treated ingots from four of these heats seriously affects the value of the comparative results.

Perhaps the authors intended to base their conclusions regarding the cleanness of the aluminum-treated steels on the chemical analyses only, although this is not stated. These analyses show dirtier steel from the mold addition of

aluminum only in two heats, and it would be interesting to have the reason explained for the disagreement with the microscopic results. The latter method is preferable since the form and distribution of the inclusions are revealed by it. The statement on line 10, page 960, regarding heat 5621 seems rather inappropriate, since this heat showed not only a 40 per cent greater inclusion count for the aluminum-treated ingot, but also 44 per cent greater total oxide content by analysis. Is not this a sufficient difference to constitute a noticeable effect?

In regard to deoxidizers other than aluminum, the statement about tiny oxide particles from titanium is at variance with general experience, and some evidence in support of such a statement would seem to be required. It is a well-known fact that titanium oxide is used as a flux in ceramic practice, while alumina and zirconia are not fluxes but serve as very efficient refractories. The writer has never seen any characteristic titanium oxide inclusions in low or medium carbon steels, nor any illustration or description of oxide inclusions due to titanium in such steels, and if the authors have discovered them a description would be of interest, and the method of use of the titanium for forming them should be stated. Alumina inclusions are on the other hand well known, and steel treated with aluminum can almost always be identified by their presence. The statement that titanium gives rise to oxide particles like alumina is contrary to past experience; in fact titanium in the form of TAM Ferro Carbon-Titanium has been and is now being used to replace a major portion of the aluminum for grain-size control for the express purpose of securing a more uniform product with fewer inclusions of alumina and other oxides or slag. The further tests made by the authors along this line should show some interesting results.

Written Discussion: By H. W. McQuaid, research metallurgist, Republic Steel Corp., Massillon, Ohio.

This study of grain-size control of plain carbon open-hearth steel is one which will be of value to the metallurgists interested in grain-size control and grain-size specifications. The authors are to be congratulated on the clarity of the presentation and the detail with which the investigation has been carried out.

It should be noted by the metallurgists who specify grain-size on plain carbon steel that in no case was the grain-size finer than 7 or coarser than 2. This should indicate that where a grain-size specification for plain carbon steels calls for a range of 6 to 8, that the chances of getting any heats finer than 7 are very remote. It should also be noted that no evidence is given that intermediate grain-size ranges are possible to obtain, and hence specifications calling for 4 to 6 grain-size are apparently very difficult to produce and should not be specified without very careful consideration of the reason for such a range.

It should also be noted that the aluminum additions required, together with the increased silicon and manganese content, result in a steel which pipes more deeply and hence the yield being less the price should be more.

This paper would lead one to believe that by a simple aluminum addition to the ladle or to the ladle and molds, that the production of either fine-grained or coarse-grained steels in the plain carbon analyses is a very simple everyday matter. In my opinion, the proper control of grain-size goes considerably be-

yond this point since we are interested not so much in grain-size control as we are in obtaining certain characteristics of which the grain-size is an indicator.

The problem of adding aluminum to the low carbon, low silicon heats is one which requires considerable care, since the results to be obtained by aluminum in this type of steels are affected not so much by the final silicon content or the final manganese content as they are by the degree of oxidation of the bath prior to the addition of manganese or silicon.

The control of grain-size, particularly in the alloy type of steel, has been practiced for several years, and one of the pioneers in this development is T. W. Hardy, who studied and used the effects of grain-size control approximately ten years ago.

The effect of aluminum additions on plain carbon carburizing grades of steel as it affected toughness, hardenability, distortion, etc., was considered and discussed and actually utilized more than 10 years ago. At this time the development was considered more or less of a secret, but the demand of the steel user for grain-size control soon led to the general knowledge of the effect of the aluminum addition.

The control of the grain-size on large open-hearth furnaces on a commercial scale is not quite such a simple matter as this paper might lead one to believe and involves a very serious study of the factors which affect deoxidation. The statement that grain-size is the principal factor in "body" may be true, but I have seen coarse grain, very abnormal Swedish steels which were noted for their body, and at the same time I have seen very fine grain and very normal steels in the alloy type which were notable for their lack of body. The word "body" here being used to indicate a certain combination of ductility, hardness, soundness and excellence of fracture, which is the ideal of every steel maker for his quality heat treating steels.

The statement that grain-size and grain-size effects are due to the obstructing action of nonmetallics should be used with caution, especially when we are talking of the McQuaid-Ehn test. We must be careful to distinguish between grain growth effects and grain-size effects, since a steel may be fine-grained at one temperature for heat treating and be coarse-grained at a temperature not over 100 degrees Fahr. higher, and when we are talking about grain-size we should be careful to state that we mean the grain-size at 1700 degrees Fahr. which may or may not be the grain-size which we desire at the temperature which would be used in processing.

Authors' Reply

We wish to thank Mr. Comstock for his careful analysis of the data on the inclusion counts. Although this indicates that the maximum additions of aluminum in the mold raised the average inclusion counts by 14 per cent above the comparison samples with no aluminum in the mold, we are still of the opinion that the aluminum additions did not adversely affect the cleanness of the steel. Owing to the very wide variation in the inclusion counts (covering a range between 10 and 108 as shown in Mr. Comstock's table) between the different heats, we do not feel that the calculated increase of 14 per cent is of

significance. On the contrary, the significant fact, it seems to us, is that in the four heats designated in the paper as the cleanest, the mold additions of aluminum showed no tendency whatever to raise the inclusion count. This is indicated below.

Heat Number	Total Oxide Count: No Mold Addition	Total Oxide Count: Maximum Al in Mold
15125	10	13
21781	36	11
23949	24	18
23950	19	22
Average	22	16

In no case did the maximum mold additions of aluminum (8 ounces per ton) in these cleaner heats raise the inclusion count to a value comparable with the rest of the heats. The latter gave an average inclusion count of about 55 as against about 20 for the cleaner heats. Thus the general cleanliness rating of the two groups of heats did not appear to be materially affected by the mold additions.

Moreover, as emphasized in the paper, we do not in general recommend producing fine grain by mold additions of aluminum but by ladle additions. It will be observed that the group of four cleaner heats was fine-grained without any mold additions, having had a ladle addition of aluminum of 16 ounces per ton. In these heats the ladle additions of aluminum required to cause fine grain evidently had no adverse effect on the cleanliness. Heat 22523, which was also fine-grained as poured from the ladle, having a ladle addition of aluminum of 16 ounces per ton, was in the group with higher inclusion count. It does not appear, however, that the aluminum addition was the cause of the high inclusion count. The other heats with considerably lower ladle additions of aluminum and coarse-grained as poured from the ladle were also in the higher inclusion-count group. Thus there appears to be no indication in the data in the paper that the moderate ladle additions of aluminum required to give fine grain adversely affected the cleanness of the steel.

It should be stated that although very small isolated inclusions were not counted all inclusions and segregates of fine inclusions over ½ inch long at 100 magnification were counted, according to the method cited in the paper. Counts on some of the specimens from the tops of the ingots were omitted because of piping—most of the experimental ingots being poured into ordinary molds instead of the hot-top molds really required after the aluminum additions. That these omissions affected the general conclusions drawn seems doubtful, but the point might be checked in further tests.

Since titanium is a strong deoxidizer, there is little doubt that it forms an oxide which may occur in steel as an inclusion; as is well known it also forms a characteristic copper-colored cubic nitride frequently observed as an inclusion. The oxide inclusion does not appear to have been definitely identified, although mention of rounded violet-colored inclusions which appeared to be titanium oxide has been made.*

*G. K. Burgess and W. Quick, "A Comparison of the Deoxidation Effects of Titanium and Silicon on the Properties of Rail Steel," Bureau of Standards Technologic Paper No. 241, Oct., 1923.

In reply to Mr. Woodhall's discussion, it should be emphasized that he is referring to steel castings. In castings it is well known that aluminum additions may cause low ductility. However, as has been pointed out first by Comstock and later by Sims, the low ductility is not due to contamination by aluminum inclusions but to the effect of strong deoxidation in causing the formation of sulphide films or sprays in the outlines of the dendrites of the cast steel. This, no doubt, may also occur in ingots, but in rolled steel the sulphide sprays are elongated and thus rendered fairly harmless.

It is true that an aluminum-treated fine-grained steel does not harden so readily or deeply as a coarse-grained steel, but as has been indicated in the paper the relative advantages and disadvantages of the two types of steel must be weighed for any particular use.

Replying to Mr. McQuaid's thoughtful discussion, we would state that only the essentials of grain-size control were described for the sake of simplicity and so as not to encumber the paper with a lengthy discussion of slag control. Presumably every competent open-hearth shop sets up a fairly standard practice, and we wished to indicate how grain-size control might be superposed on such a practice without unduly changing it. Where adequate slag control is used, the manganese and silicon content of the finished steel should be a fairly good index of the degree of deoxidation. In general, the more thoroughly the bath is deoxidized before the addition of manganese and silicon and the less opportunity for oxidation thereafter, the less aluminum will be required to give fine grain.

The association of "body" with actual grain-size seems so significant that any instances where this relationship does not appear to hold should be looked upon as the exceptions which prove the rule. Very probably a careful metallographic examination of such apparent exceptions would indicate that they are explainable on the basis of well-known structural characteristics.

THE P-F CHARACTERISTIC OF STEEL

By B. F. Shepherd

Abstract

This work describes a comparatively simple and accurate method of determining differences between various heats of steel of similar chemical analysis. This difference has been found to be a definite property and has been termed the "P-F Characteristic."

By use of the test, heats of steel can be classified as definite types which have a definite relation to melting practice, ingot yield, hardening sensitivity and physical properties of the hardened part.

The test has been confined principally to tool steels but the adaptability to carburizing and other types has been indicated. The results of this test are directly interpretable in terms of manufacturing practice.

During the development of the test a set of martensitic fracture standards was developed. An earnest recommendation is made that this set of standards be universally adopted to avoid the confusion which would necessarily arise from the development of standards in individual plants.

FOR many years steel makers and steel users have expressed preferences for certain brands of steel claiming superiority due to some inherent property vaguely described as "body," "timbre," "personality," etc. Considerable controversy has arisen as to whether such major differences in quality could exist between steels of apparently similar chemical composition.

In 1922 McQuaid and Ehn found that certain heats of steel had a tendency to harden in spots after being carburized and quenched from the hardening temperature. They were able to isolate such heats of steel by the use of the McQuaid-Ehn test, which consisted in noting the condition of the hypereutectoid zone after carburizing at a high temperature. This immediately resulted in a demand for the so-called "normal" type and since that time there has been very little "abnormal" carburizing steel manufactured.

A paper presented as part of the Grain-Size Symposium of the Sixteenth Annual Convention of the Society held in New York City the week of October 1, 1934. The author, B. F. Shepherd, president of the Society, is manager of the Rock Drill Division of the Ingersoll-Rand Co., Phillipsburg, New Jersey. Manuscript received July 5, 1934.

At some later date it was found that the grain-size of the hypereutectoid zone of steel given the McQuaid-Ehn test gave an index of the response of that steel to heat treating operations. In general, the so-called coarse-grained steels hardened harder and deeper, and had much lower impact strength and better machinability.

In 1926 we began distinguishing the differences between tool steel by means of the Shepherd Hardenability test. This test was made at the ordinary range of hardening temperatures but elevated temperatures were included later. A considerable difference was found in the behavior of various heats (of the same hardenability at 1420 degrees Fahr.) after quenching from elevated temperatures. This difference consisted in a change in the hardenability and also in the martensitic fracture grain-size.

Classification of fracture grain-size was extremely difficult until a set of martensitic fracture grain standards was developed. It was then possible to describe fracture grain-size tests numerically. These standards are shown in Fig. 1 with corresponding microstructures in Fig. 2.

The classification of fractures requires a great deal of personal skill. Even though a set of fracture standards is available, proper technique must be used in interpreting values. The following suggestions are made as an aid to such interpretations:

1. Any fracture determinations should be an agreement of two qualified observers. Disagreement by more than ½ fracture number shall result in arbitration by a third observer and the apparent true fracture number recorded.

2. To assist in eliminating personal error, classifications are made to ¼ number.

3. Observers are to be qualified by passing the following tests on a set of approximately twenty samples of various fracture classifications.

   a. Ability to check their own classifications on different days to within ¼ fracture number on 90 per cent of the samples.

   b. Apparent true values should be established on approximately twenty samples by two or more observers passing Test No. 1. To qualify, at least 80 per cent of these samples must be classified within ¼ fracture number.

A representative qualification test is shown in Table I.

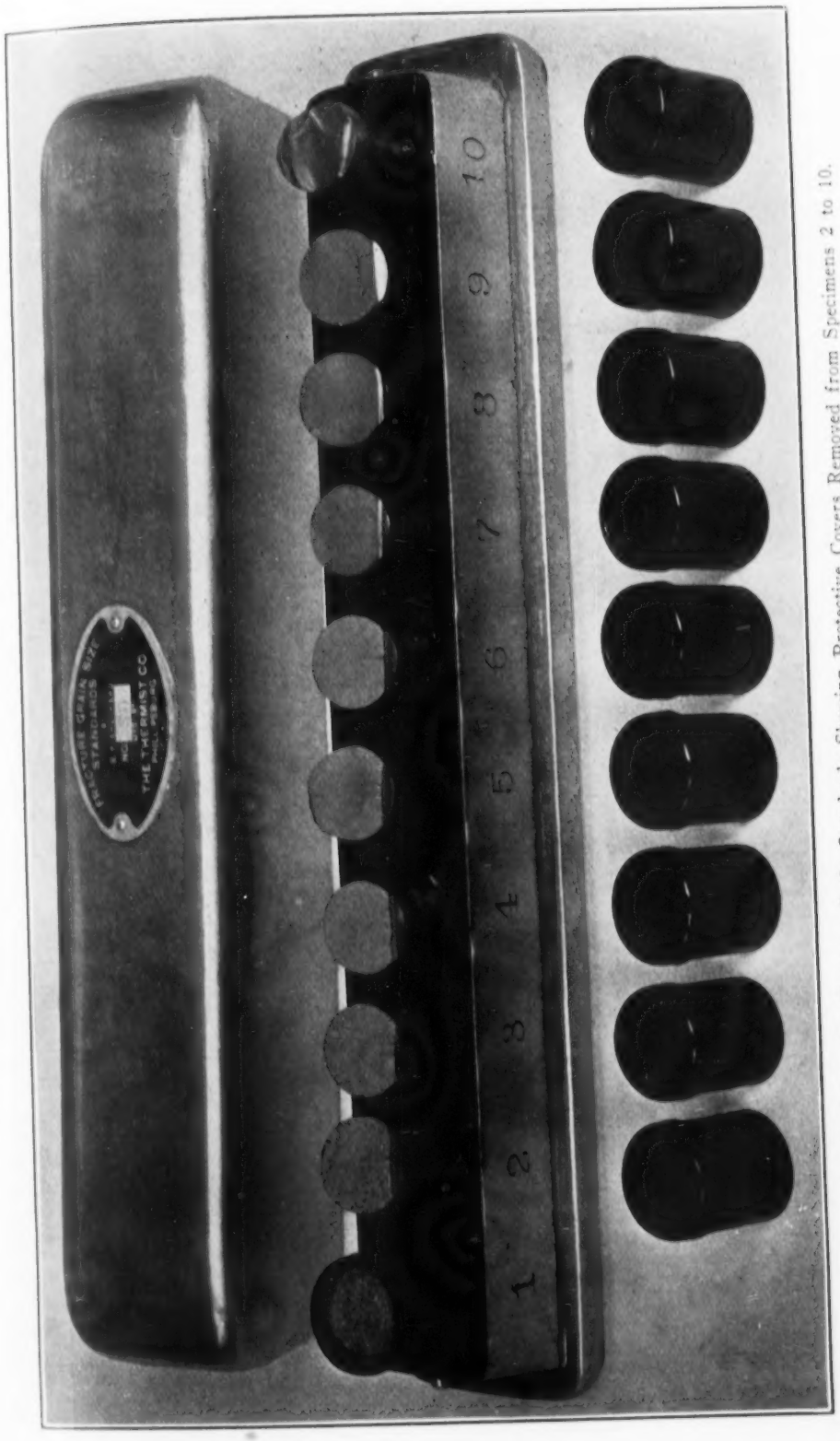


Fig. 1—Set of Martensitic Fracture Grain Standards Showing Protective Covers Removed from Specimens 2 to 10.

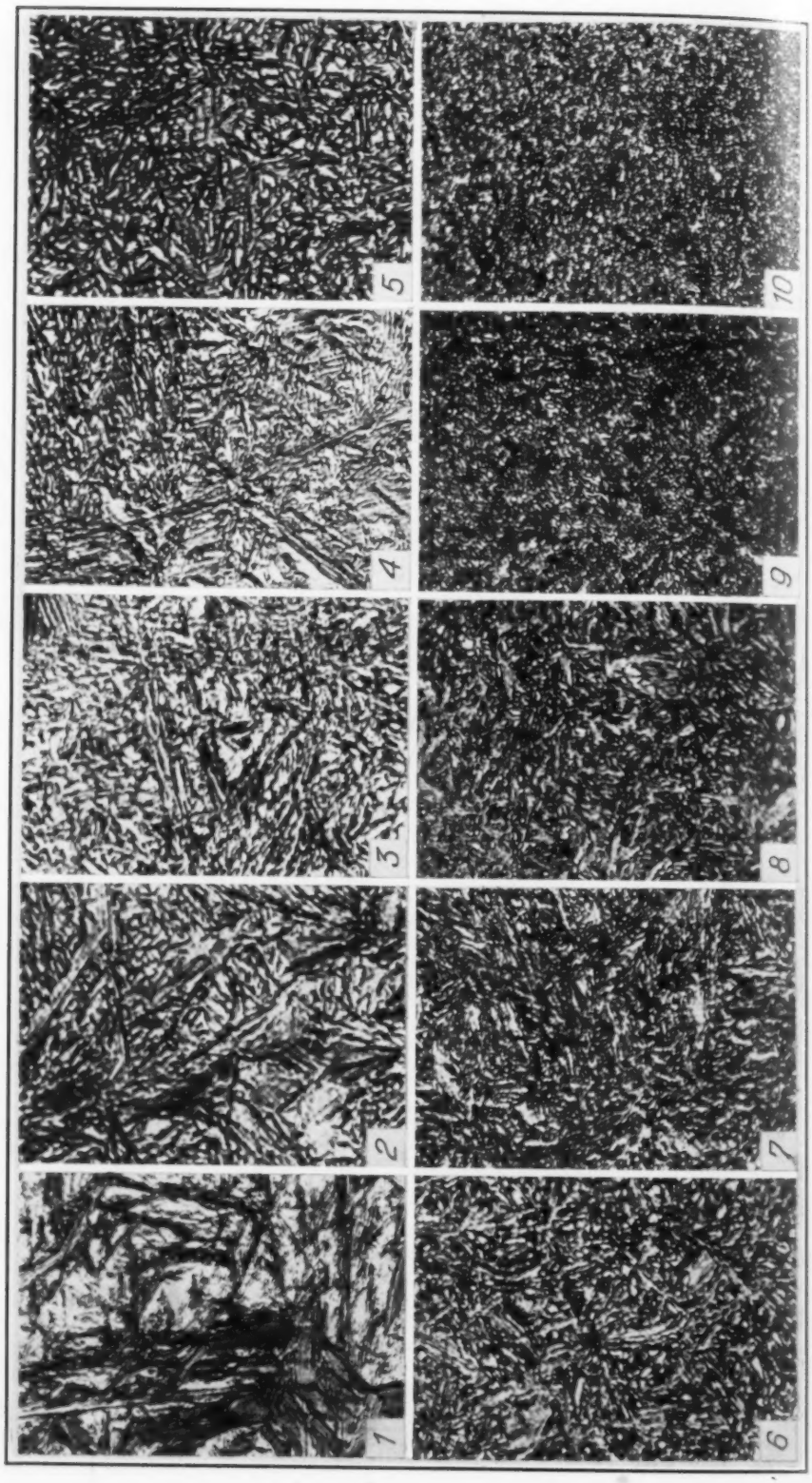


Fig. 2—Photomicrographs of the Structures of Fracture Grain Standards Shown in Fig. 1.

Table I
Qualification Test for Fracture Observer

Sample No.	Apparent True Fracture No.	Observer — Variation from Apparent True Number			
		A	B	C	D
1	8	+¼	0	+¼	−1
2	4½	0	0	0	−1
3	7½	0	+¼	0	−1
4	8¼	0	−1	0	−2¼
5	3½	+¼	+¼	−¼	−¼
6	8½	0	+¼	0	−½
7	5	0	0	+¼	0
8	8¼	0	0	+½	0
9	3	0	+¼	0	0
10	8¾	−¼	−¼	+¼	0
11	8¾	+¼	−¼	0	−¼
12	7½	0	0	+¼	−2
13	4¼	0	0	+½	+¼
14	8¾	0	−¼	+¼	−½
15	3¾	+¼	0	−¼	0
16	8¼	0	−¾	−¼	−¾
17	3¼	0	0	−¼	+½
18	8¾	−¼	0	−¼	−¼
19	8	0	+½	0	+½
20	3	0	0	−¼	0
	Grade	100	85	90	50

Penetration (P) — Fracture (F) Test Method

1. *Preparation of Test Pieces*

(a) It is most important to obtain metal that is representative of the heat of material to be tested. Therefore, stock whose diameter is greater than 1¼ inch is reduced in size to 1-inch round (approximately) by suitable forging. Approximately 14 inches of material is required in lengths not less than 3 inches. When possible, the original marking is allowed to remain untouched in the forging operation, thus avoiding mixed samples.

(b) The sample material may be annealed before machining (1325 degrees Fahr. air or 1450 degrees Fahr. furnace cool, as convenient).

(c) Four specimens are machined ¾-inch round, plus or minus 0.001 inch and 3 inches long. Each piece is stamped with a suitable code or number for identification. The pieces are also stamped on both ends, 1450, 1500, 1550 and 1600 respectively, one number to each piece.

2. *Normalizing*

All samples are given an oil-quench from 1600 degrees Fahr. after holding the pieces at temperature a minimum of 40 minutes after heating through.

3. *Hardening*

(a) Heating—An electric furnace should be used for heating.

Fig. 2—Photomicrographs of the Structures of Fracture Grain Standards Shown in Fig. 1.

Fig. 3—Showing Method of Quenching Specimens in a Standard Vertical Jig.

The heating rate should be similar to or the equivalent of that obtained in a 11 KW top and bottom element electric furnace with a hearth size of 12 by 30 inches, and not more than four samples shall be heated at one time. The samples shall be placed erect in the furnace, and so located that all samples are heated at the same rate. Temperatures of 1450, 1500, 1550 and 1600 degrees Fahr., respectively, shall be used, the samples having been so identified previously. The samples are in the furnace for 30 minutes for each temperature.

Accurate control of the time-cycle and temperature must be maintained. This is extremely important because penetration of hardness and fracture grain-size tend to increase with time and temperature on all except very stable heats.

(b) QUENCHING—The samples shall be quenched in a standard vertical jig (2-inch pipe overflow, Fig. 3), flushed with 10 per cent brine at room temperature.

Fig. 3 shows a specimen in the tongs being placed in the spray. The tongs have a shoulder which enables the specimen to be placed easily in the center of the spray. The P-F test section (Fig. 5) shows the uniformity of distribution of hardness along the length. The soft spot has been caused by the tongs.

4. *Fracture Grain-Size*

(a) BREAKING—The samples are notched halfway between

ends with a thin ($\frac{1}{16}$ to $\frac{3}{32}$ inch) friction wheel, to a depth not over $\frac{1}{16}$ inch. The fracture is obtained preferably by transverse impact because transverse static loading to rupture tends to pull or otherwise distort the grain.

(b) The grain-size is determined by comparing and matching the case or hardened zone of one of the broken halves with the fracture grain-size standards, and shall be expressed as the number or quarter number nearest the corresponding standard sample. Determination of the grain-size number is made according to the standard method.

5. *Penetration*

The penetration tests are made on the second half of the hardened and fractured sample. A smooth cross section surface is prepared not less than 1 inch from the original end of the hardened

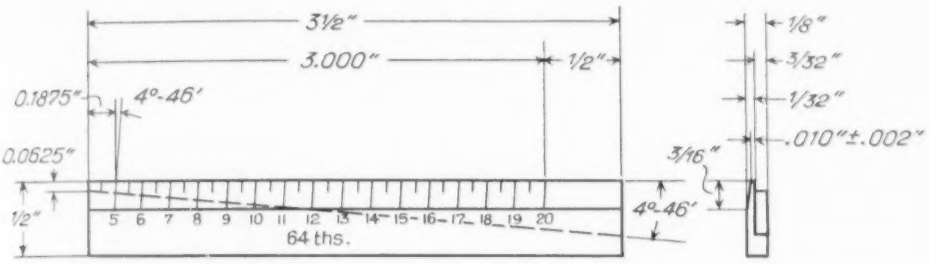


Fig. 4—Gage for Measuring P-Value.

sample. Final finish on the cross section surface is on an "0" or finer abrasive. The polished surface is etched for 3 minutes in a 50-50 hydrochloric acid and water at 180 degrees Fahr.

The hardness penetration is measured, macroscopically, to the nearest half of $\frac{1}{64}$ inch and shall be expressed as the numerator of such a fraction. A convenient scale is shown in Fig. 4.

6. *Designation*

The P-F characteristic is recorded as 8 numbers. The first four numbers represent the penetration in $\frac{1}{64}$ths of an inch, and the last four numbers the fracture grain-size. The first number of each group represents the P-F result for a temperature of 1450 degrees Fahr., the second number of each group represents the P-F for a temperature of 1500 degrees Fahr. and so on.

The characteristic then appears in the following form:

5. 6. 8. 12. / 9. 8. 6. 4.

When all the results do not concur with whole numbers, the characteristic may appear as follows:

$$5\frac{1}{2}.\ 6.\ 8\frac{1}{2}.\ 13.\ /\ 8\frac{3}{4}.\ 8\frac{1}{4}.\ 7\frac{1}{2}.\ 5.$$

In developing the test method electric furnaces were compared against the lead pot method of heating. The following typical heat-

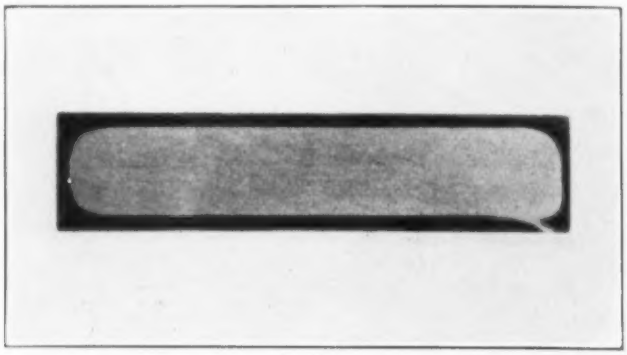


Fig. 5—P-F Test Section Showing the Uniformity of Distribution of Hardness Along the Length. The Soft Spot Was Caused by the Tongs.

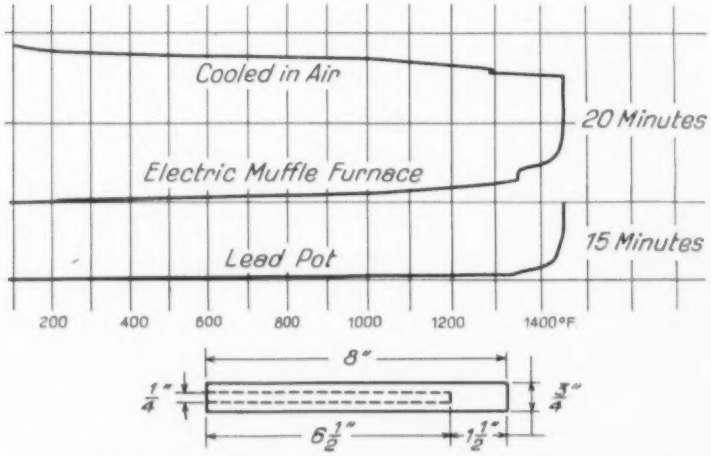


Fig. 6—Comparison of Rate of Heating a ¾-Inch Round Specimen of Electric Furnace 1.05 Per Cent Carbon Tool Steel in an Electric Muffle Furnace and Lead Pot.

ing curves (Fig. 6) are characteristic of the results. A lead pot takes almost as long to heat these small samples to temperature as an electric furnace. The initial heating is very rapid and if a piece ¾-inch round was quenched after five minutes in the lead pot it would harden, whereas in the electric furnace it would be just about in the transformation range.

In comparing oil-treated versus annealed samples it was ob-

served that the oil-treated samples apparently heated much more rapidly than the annealed samples. Investigation with test specimens showed that this was due to a lowering of the critical point on the oil-treated samples as shown in Fig. 7. Apparently the carbides which have been made very small and uniform by the oil-treating operation are much more readily soluble. For a long, slow heating the critical points are the same. Critical point determinations which

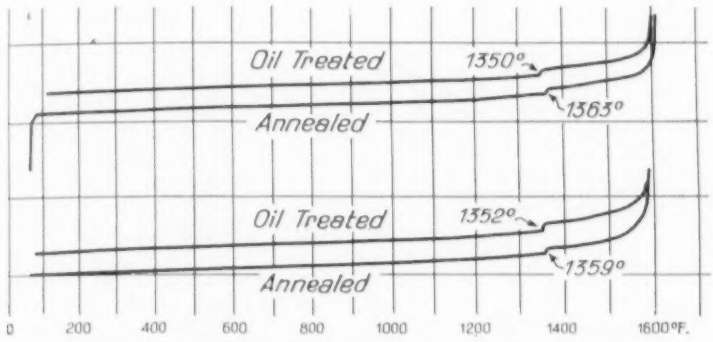


Fig. 7—Influence of Structure on the Critical Point of ¾-Inch Round 1.05 Per Cent Carbon Tool Steel.

have been made by heating rapidly and have not been accompanied by a statement as to the previous microstructure, should not be given credit for strict accuracy.

The P and F values should not be considered separately when considering the P-F characteristic. In order to illustrate this, four heats are shown in Table II. On the first two heats, the P characteristic is practically the same, while there is a marked difference in the F characteristic. On the second and third heats the F characteristic is about the same but there is a marked difference in the P characteristic. The fourth heat has a very stable but coarse F characteristic.

It is not always desirable to give the full P-F characteristic

Table II
Difference in P and F Values Between Three Basic Open-Hearth Heats

Heat No.	Penetration 1450	Penetration 1500	Penetration 1550	Penetration 1600	Fracture 1450	Fracture 1500	Fracture 1550	Fracture 1600
1D091	8	11	24	24	8¼	7¼	3¾	3½
30420	9	11	24	24	8¼	7¼	6¼	5½
5D330	7	8	9	13	8¼	7¾	7¼	5½
CPX	12	16	16	24	4½	4½	4½	4¼

Chemical Analysis

Heat No.	C	Mn	P	S	Si
1D091	1.01	0.27	0.015	0.021	0.15
30420	0.98	0.29	0.016	0.028	0.20
5D330	1.06	0.27	0.015	0.020	0.15
CPX	0.80	0.31	0.023	0.017	0.15

value for a heat in discussion, and the heats are occasionally referred to as types. For tool steel we used arbitrary classifications as follows:

Type 1 was used to designate a steel which was stable or insensitive to increase in hardening temperature, both with regard to P and F characteristic and Type 3 as being directly opposite. Type 2 was intermediate.

Other types are classified as fractions. A Type 2-1, for example, is intermediate in P characteristic and stable with regard to its F characteristic; whereas a Type 1-3 would be stable with regard to penetration and very unstable with regard to fracture.

Type	P	F
1.	$\frac{3}{64}$ inch or less change.	Not more than one number.
2.	More than $\frac{3}{64}$ inch, but does not harden through.	More than one (1) number; less than two (2) numbers.
3.	Hardens through at 1600 Fahr.	More than 1¾ number change.

Incidentally, when a number of heats of various types were compared with the yield of finished bars passing fracture inspection, all of the points fell on the line; Type 1 having a high yield and Type 3 a comparatively low yield. Table III illustrates P-F types in electric furnace 1.05 per cent carbon steel based on the differences between 1450 and 1600 degrees Fahr. values.

Table III

P-F Types in Electric Furnace 1.05 Per Cent Carbon Tool Steel

		Penetration Degrees Fahr.				Fracture Degrees Fahr.			
Heat No.	Type	1450	1500	1550	1600	1450	1500	1550	1600
H5194	1	6	6	7	8	9	9	9	8½
6101	1/3	5	5½	6	7½	9	8½	6¼	4½
H4705	2/1	6	7	8	12	8¾	8½	8¾	8¼
X6892	2	5½	6½	8	14	8¾	8½	8¼	7¼
H5183	2/3	5	5½	7	13	8¾	8½	8¼	5
29983	3	7	8	9	24	8½	8¼	6¼	4
W5294	3	7	8	10	24	8¾	8¼	7	6¼

		Chemical Analysis						
Heat No.	Type	C	Mn	P	S	Si	Ni	Cr
H5194	1	1.07	0.21	0.010	0.015	0.17	0.09	0.03
6101	1/3	1.07	0.20	0.013	0.013	0.20	0.025	0.01
H4705	2/1	1.05	0.21	0.010	0.014	0.19	0.10	0.05
X6892	2	1.06	0.23	0.014	0.013	0.25	0.09	0.03
H5183	2/3	1.05	0.25	0.011	0.017	0.17	0.08	0.02
29983	3	1.06	0.07	0.011	0.011	0.14	0.06	0.05
W5294	3	1.05	0.21	0.011	0.015	0.24	0.06	0.03

In standardizing the test, heating times of 30, 45 and 60 minutes were used with typical results shown in Table IV. It is interesting to note that Specification 76 is so stable that it makes practically no difference in the penetration or fracture value, whether 1450 or 1600 degrees Fahr. is used for hardening.

Table IV

Effect of Heating Time on Penetration-Fracture Values of Several Types of Steel

Steel	Heating Time Minutes	Penetration Degrees Fahr. 1450	1500	1550	1600	Fracture Degrees Fahr. 1450	1500	1550	1600
ED	30	5	5½	6	6¼	9	8¾	8½	8¼
Type 1	45	5	5½	6½	7	8¾	8½	8¾	8¼
Spec 1	60	5	5½	7	7½	8¾	8¼	8	7¾
BM	30	7	8	9	17	9	8½	8	7½
Type 2	45	7	8	9½	T	8¾	8½	8	6½
Spec 1	60	7	8	10	T	8¾	8½	7½	6¼
DN	30	7	9	13	T	9	8¾	8¼	6½
Type 3	45	7½	10	14	24	8¾	8¼	7¾	5¾
Spec 1	60	7½	11	24	24	8¾	8	7¾	5½
2D	30	7½	9	12	24	8½	8	7½	6½
Spec 332	45	7½	9	14	24	8½	8	7	5¾
Type 3	60	8	9½	16	24	8½	8	6½	5½
2H	30	14	14	14	15	8	8	7¾	7¾
Type 1	45	14	14	15	15	8	8	7¾	7¾
Spec 76	60	14	15	15	15	8	8	7¾	7¾

Our previous tests and experience with regard to hardenability test specimens would indicate that the effect of spray or hand agitation would have little effect upon the results. This was borne out by the tests which were made on various types of steel as shown in Table V.

Table V

Effect of Spray —S— Immersion Vs. Hand —H— Agitation in Quenching Penetration-Fracture Test Pieces of Electric Furnace 1.05 Per Cent Carbon Tool Steel

Heat No.	Quench	Type	Penetration Degrees Fahr. 1450	1500	1550	1600	Fracture Degrees Fahr. 1450	1500	1550	1600
30127	H	1	5			6	9			8½
30127	S		5			6	9			8½
W5907	H	2	7			17	9			7½
W5907	S		7			17½	8¾			7¾
W5294	H	3	7			24	8¾			6¼
W5294	S		7			24	8¾			6
5760	H	1/3	5	5½	6½	7	8¼	7¾	5	4½
5760	S		5	5½	6½	7½	8½	7¾	5	4½
5877	H	2/3	5½	6	8	10	8½	8	5½	4¾
5877	S		5½	6	8	10	8½	8	5½	4¾
H4706	H	2/1	5			9	9			8¼
H4706	S		5			9	8¾			8

Segregation has an influence on hardenability. We find dumbbell fractures in hardened test pieces occasionally and these are characteristic of the heat, showing a center condition which increases the depth of hardening. A standard ⅜-inch slab test piece and two test pieces from different heats showing varying degrees of dumbbell fracture are shown in Fig. 8.

Hardenability is a heat characteristic. A heat of electric furnace steel was explored by taking the first, middle and last ingots and mak-

ing a Shepherd hardenability test from the top, middle and bottom of these ingots. Fig. 9 is self-explanatory.

An investigation of some defective steel showed that a reaction had taken place in the ladle which had progressed sufficiently far during the casting of the last ingot to cause a considerable difference in the billet tests, the etched bars and the P-F characteristic.

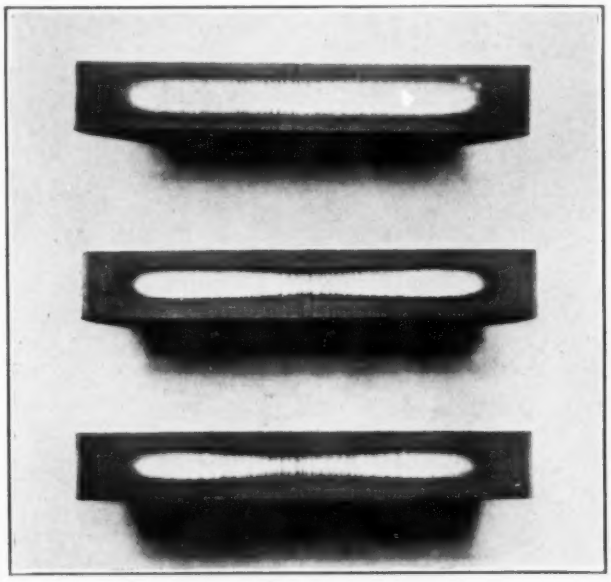


Fig. 8—A Standard ⅜-Inch Slab Test Piece and Two Test Pieces from Different Heats Showing Varying Degrees of Dumb-bell Fracture.

This indicated that the P-F characteristic is due to the melting reactions and not to the ingot size, temperature of pouring, etc. Figs. 10, 11 and 12 are self-explanatory.

An electric furnace heat and an open-hearth heat were surveyed to determine whether the P-F characteristic was an inherent property of the heat itself. Tables VI and VII, Figs. 13 and 14, show that this characteristic is a function of the heat. It is evident, however, from Figs. 10, 11 and 12 that tests on both the first and on the last ingot cast are necessary for an assurance of heat uniformity.

Hardness surveys are often made of specimens and the results assumed to be independent of the prehardening structure and typical of steel of that particular chemical analysis. The following graphs, Figs. 15, 16, 17 and 18, show that the P-F characteristic and preliminary structure have a great influence upon the shape of these curves.

Table VI

Exploration of Heat 31196—Spec. 1 for Uniformity of Penetration-Fracture Characteristic. Heat was from 8-Ton Basic Electric Furnace Top-Poured in 15" Square Ingots

Penetration-Fracture Test Made on 1" Bars Hammered from Section of 4" Billet from Top of Ingot
Results Given are Deviation from Average Results

	Penetration — Penetration in 64ths — Degrees Fahr.				Fracture — Fracture Grain Size — Degrees Fahr.			
Ingot No.	1450	1500	1550	1600	1450	1500	1550	1600
Average	6	6½	7	10	8¾	8½	7	5¾
1	0	0	0	0	0	0	¼	0
2	0	0	0	0	0	−¼	−½	¼
3	0	0	½	0	0	−¼	0	0
4	0	0	0	0	0	0	¼	½
5	0	0	0	0	0	0	0	0
6	This Ingot Scrapped in Processing							
7	0	0	½	0	0	−¼	0	0
8	0	0	½	1	0	0	¼	0

Chemical Analysis of Heat

C	Mn	P	S	Si	Ni	Cr
1.06	0.19	0.010	0.014	0.26	0.05	0.05

Material furnished courtesy Carpenter Steel Co.

Table VII

Exploration of Heat 1D006—Spec. 332 for Uniformity of P-F Characteristic Heat was from 100-Ton Basic Open-Hearth Top-Poured in 24" Corrugated Ingots

P-F Test Made on 1" Bars Hammered from Section of 4" Billet from top of Ingot
Results Given are Deviation from Average Results

	Penetration — Penetration of Hardness in 64ths				Fracture — Fracture Grain Size				Chemical Analysis		
Ingot No.	1450	150(	1550	1600	1450	1500	1550	1600	C	Mn	Si
Average	7½	9	12	24	8¾	8¾	6½	5½	1.04	.30	.13
D1	0	0	−2	0	0	0	+¾	+¼	−.02	−.01	−.01
D3	0	0	0	*0	0	0	+¾	+¼	−.01	−.01	0
D4	0	−½	+3	0	0	+¼	0	0	−.01	−.01	+.01
D5	0	−½	−1	0	0	0	0	+¾	0	0	0
D6	0	0	0	0	0	+¼	0	0	−.02	0	+.01
D7	0	−½	−2	0	0	0	1	−¼	−.02	0	+.01
D8	+½	0	0	0	0	+¼	−¼	−¼	−.01	0	0
D9	0	−½	−1	0	0	+¼	−½	−½	+.01	0	0
D10	+½	0	−1	0	0	0	½	−¼	−.01	.01	−.02
D11	0	0	−1	0	−¼	−¼	−¼	−¼	0	.01	0
D12	0	0	0	0	0	0	−¼	0	0	0	0
D13	+½	0	−1	0	0	0	+¾	+¼	−.01	0	0
D14	0	0	−1	0	0	0	+½	+½	0	0	−.01
D15	+½	0	−1	0	0	−½	+1	+¼	0	0	0
D16	0	0	+1	T	0	0	−1	−½	0	0	0
D17	+½	0	0	0	0	−¼	0	−½	+.01	0	−.01
D28	+½	0	+1	0	−¼	−¼	0	0	0	0	−.02
D29	+½	0	+3	0	−¼	−¼	−¾	0	0	0	0
D30	+½	0	+2	0	−¼	−½	−¾	0	0	0	−.01
D31	0	0	0	0	−¼	0	−¼	−½	+.02	0	−.02
D32	0	0	+1	0	0	0	−¼	0	0	0	0

*Inclusion in center retarded penetration.
Material furnished courtesy Bethlehem Steel Co.

Effect of Prehardening Structure Upon the P-F Characteristic

It will be noted that practically all of the tool steel samples have been given a high temperature oil quench before the final hardening operation. This has been done to remove structural irregularities

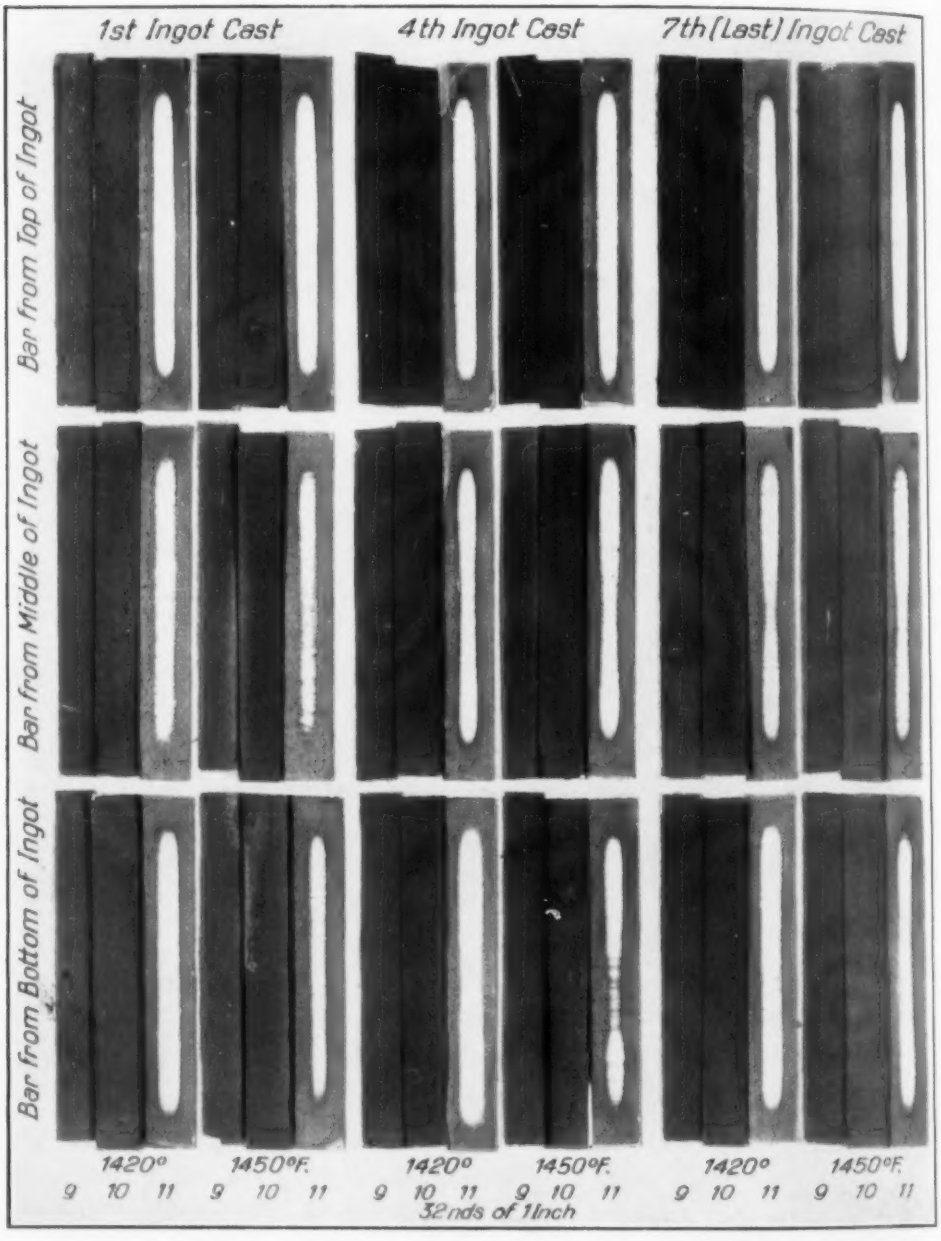


Fig. 9—Polished and Etched Specimens Showing the Uniformity of Hardenability Through the Entire Heat of Steel (Shepherd Hardenability Test). Material Courtesy Halcomb Steel Company.

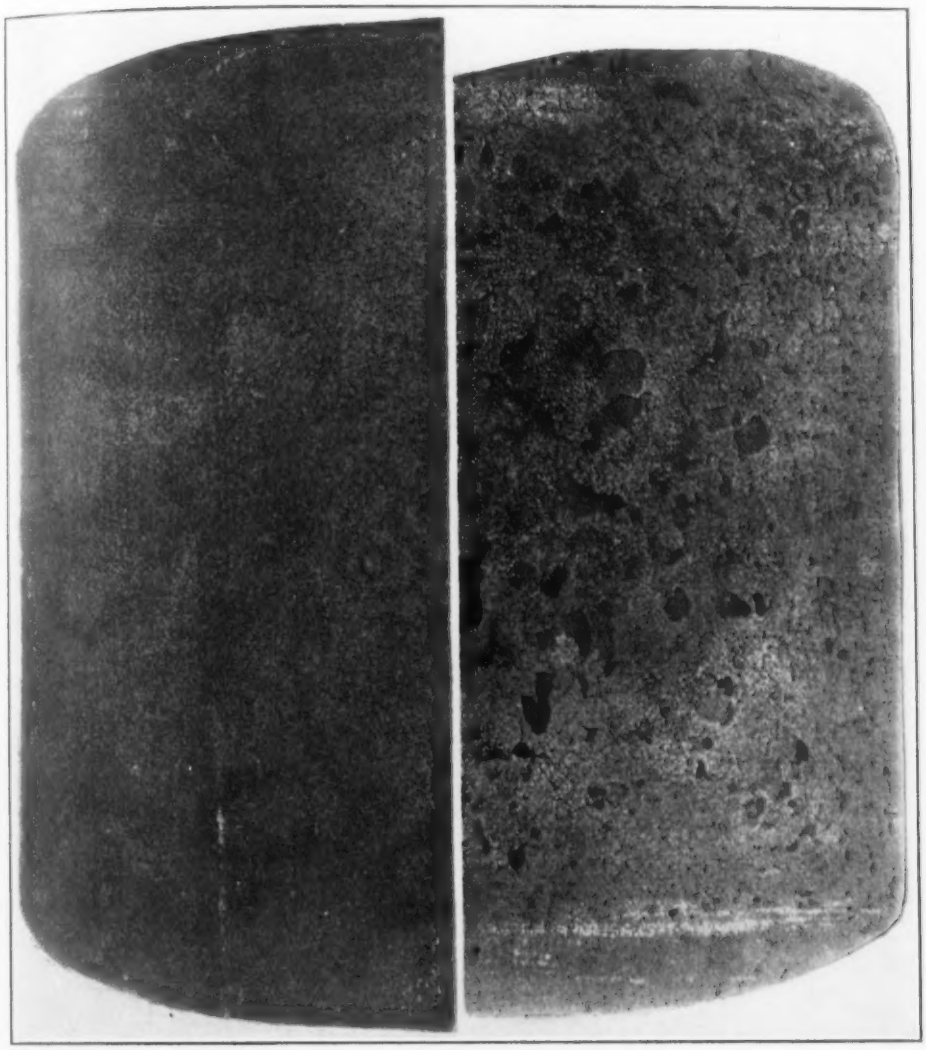


Fig. 10—Etched Sections of Half of the Cross Sections of Billets from Steel B-332. Left, Billet from Fourth Ingot. Right, Billet from Last Ingot.

and was found necessary for definite control of the P-F test. Preliminary experiments showed that the P-F characteristic of occasional heats was not sensitive to prehardening structure. The large majority of heats, however, showed a definite response to definite structure changes and this sensitivity had a definite relation to the type of steel as determined by P-F tests.

Consistent difference has been noted when the prehardening structure was changed from pearlite to sorbite. Heats of Type 1 characteristic were rather insensitive to changes in prehardening structure, while Type 3 heats showed a marked difference in the

Table VIII
P-F Characteristic Between Pearlite and Sorbitic Prehardening Structures

Fig. Number	Type	Heat Number	Structure	P-F Characteristic							
				Penetration					Fracture		
15	1	30864	Sorbite	6	7	8	9	9	8¼	8	7¼
16		30864	Pearlite	7	7	7½	8	9	8¾	8¼	8
17	3	29983	Sorbite	7	8	10	24		8¾	8½	6½ 4
18		29983	Pearlite	8	8	9	11		8¾	8¼	7½ 7¼

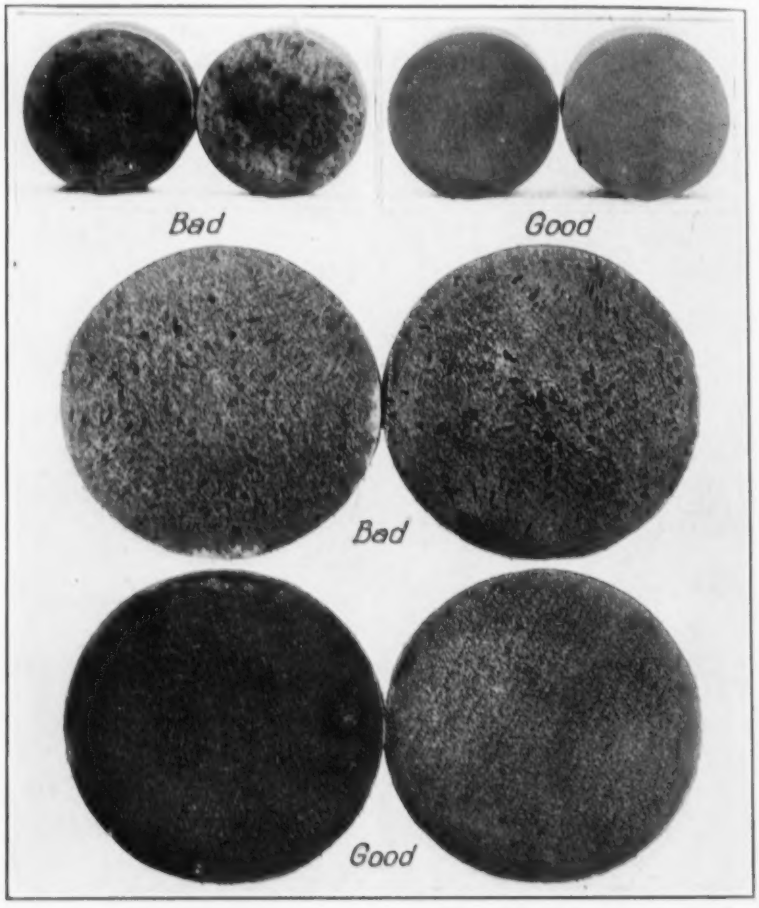


Fig. 11—Bars Rolled from Billets Shown in Fig. 10.

P-F characteristic between pearlitic and sorbitic prehardening structures. This is shown in Table VIII.

Rockwell hardness surveys show certain interesting features, see Figs. 15, 16, 17 and 18. The curves for Type 3 illustrate the points more clearly. The gradation between the martensitic and troostitic zones is much more abrupt with a sorbitic prehardening structure than with the pearlitic prehardening structure, as the gen-

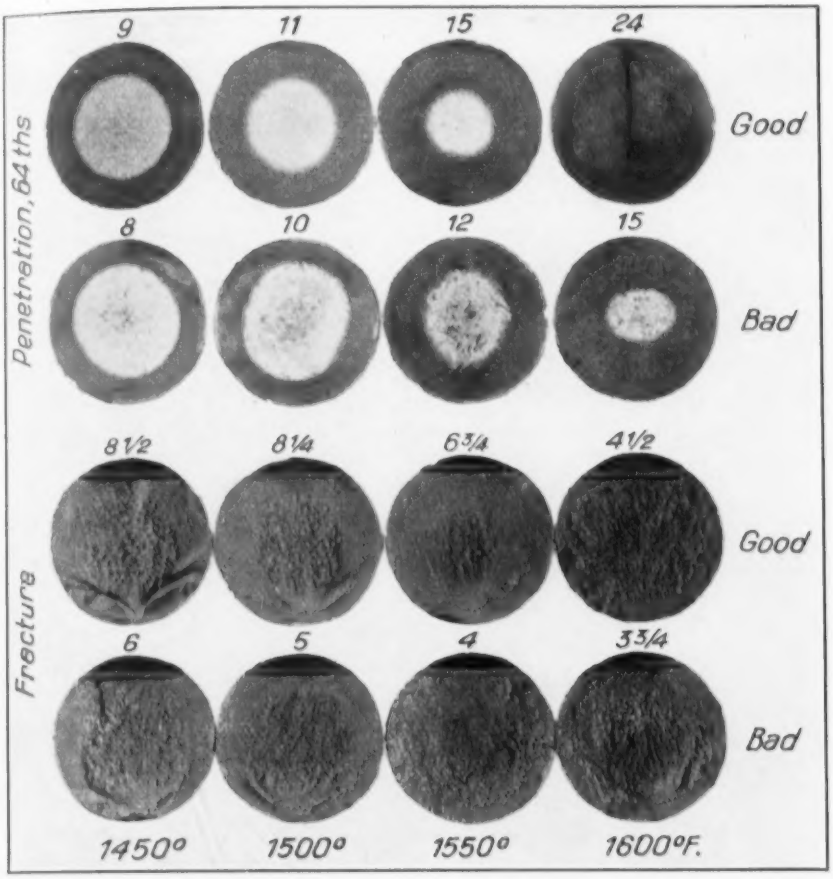


Fig. 12—Influence of Inclusions on the P-F Characteristic of Carbon Tool Steel.

eral slope of the sorbitic curves for normal hardening temperatures is steeper than the pearlitic curves. The curves also show that heating to 1600 degrees Fahr. and quenching in oil definitely reduces the penetration value over that of the same sample annealed previously at 1450 degrees Fahr. and hardened at normal temperatures. When higher temperatures (1550 and 1600 degrees Fahr.) are used, the reverse is true and the sorbitic structure tends to cause deeper penetration and coarser grain structure.

It is very difficult to obtain hardness readings close to the surface when using the cross section of P-F test pieces and a number of tests were made, using Vickers pyramid hardness testing machine. It was possible by this test to identify the first appearance of troostite, the last traces of martensite and the last traces of troostite, in the microstructure, as indicated by observation of the hardness curves.

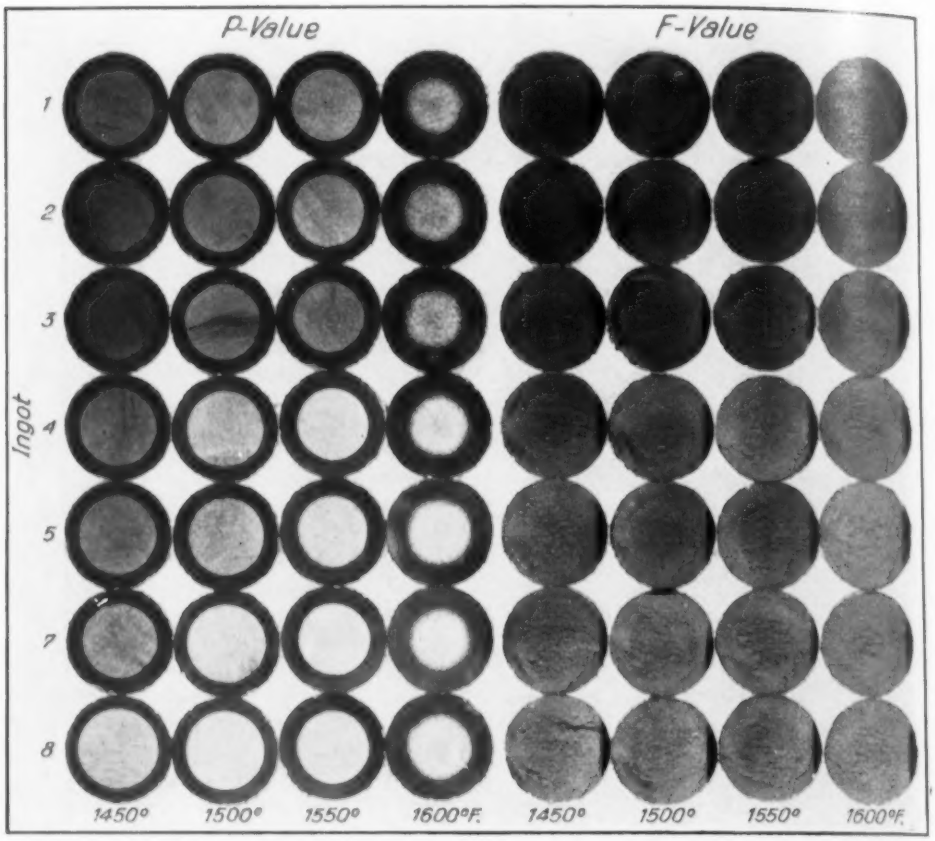


Fig. 13—Exploration of Heat 31196, Specification 1 for Uniformity of Penetration-Fracture Characteristic. (See Table VI.)

Table IX
Relationship Between Type of Steel and Uniformity of Hardened Zone

Type	Code	Heat No.	P-F Characteristic: Penetration				P-F Characteristic: Fracture			
1	BE	H-4974	6	7	8	9	9	9	9	8½
2/1	DH	X-6903	6	7	9	15	9¼	8¾	8½	8¼
2	CX	X-6892	5½	6½	8	14	8¾	8½	8¼	7¼
2/3	M	W-5129	6½	7½	8	11	9¼	9	8¼	7¼
3	CF	H-5184	7	8	9	24	9	8½	8	6¼
3	DN	H-5262	7	9	13	24	9	8¾	8¼	5

Chemical Composition

	C	Mn	P	S	Si	Ni	Cr
Average	1.06	0.23	0.013	0.014	0.20	0.08	0.04
BE	+0.02	−0.02	0.002	0	−0.02	+0.01	0
DH	−0.01	0	+0.003	−0.001	+0.02	−0.01	+0.01
CX	0	0	+0.001	−0.001	+0.05	+0.01	−0.01
M	0	−0.04	−0.001	+0.001	−0.01	−0.01	−0.01
CF	0	+0.01	−0.002	−0.004	0	−0.01	+0.01
DN	−0.12	+0.06	−0.001	+0.004	−0.02	0	+0.01

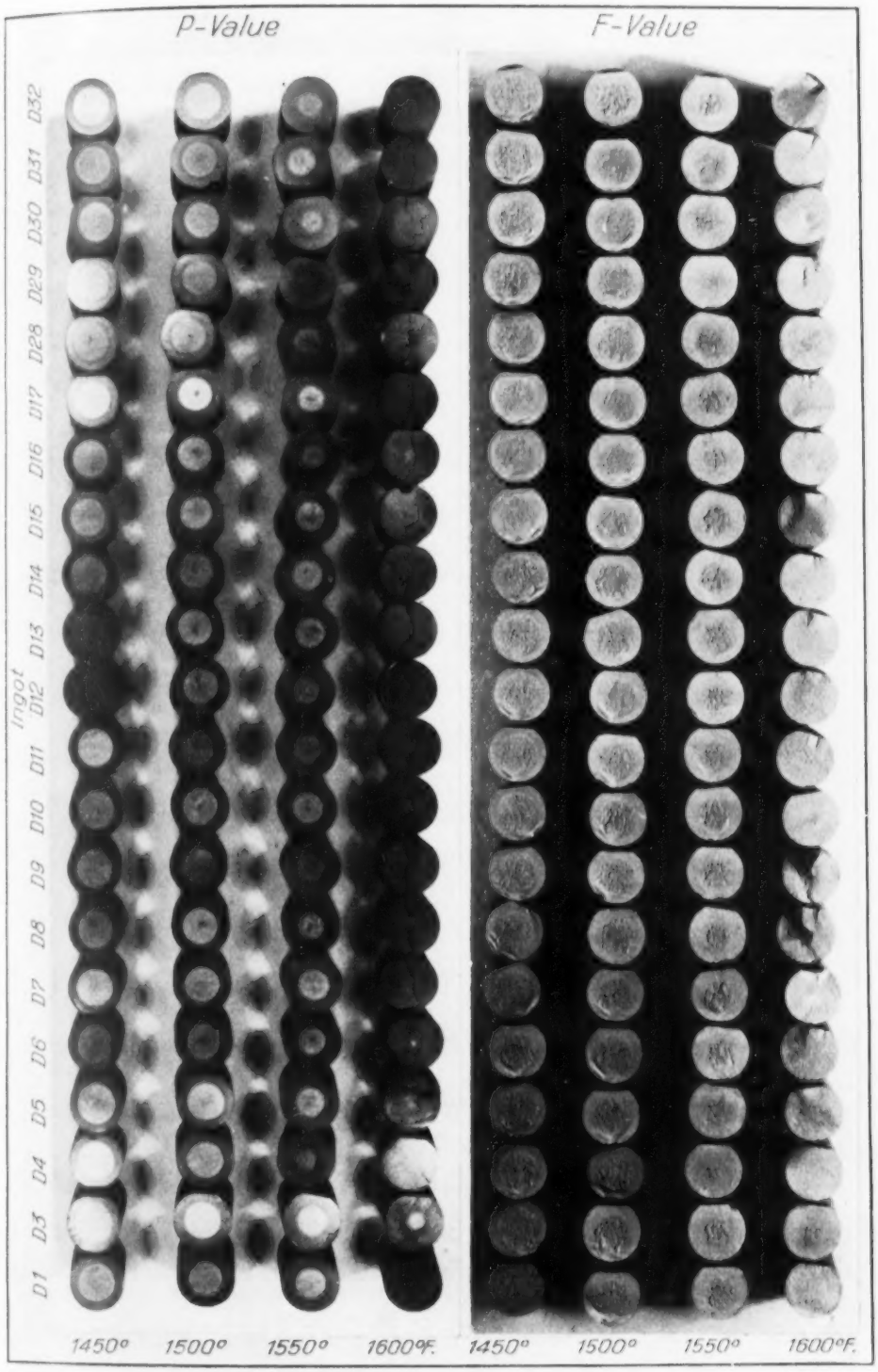


Fig. 14—Exploration of Heat 1D006; Specification 332 for Uniformity of Penetration-Fracture Characteristic. (See Table VII.)

The sensitivity of these tests shows a distinct tendency for Type 3 steels to have a more uniform hardness in the hardened zone than Type 1. There seems to be a definite relation between the type of

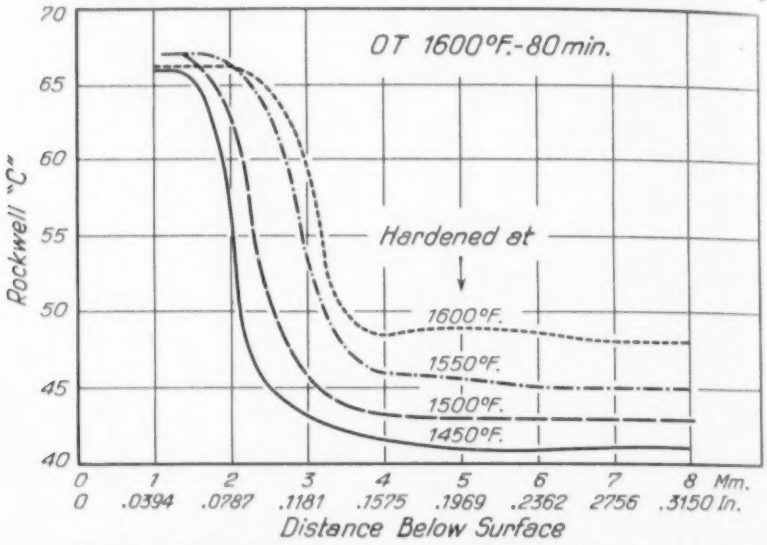


Fig. 15—Rockwell Hardness Survey of Steel X30864 Hardened as Shown. P-F Type 1, See Table VIII.

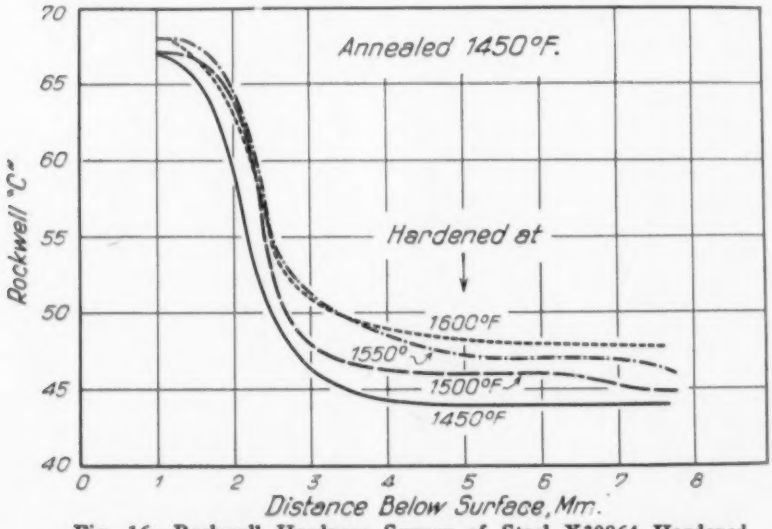


Fig. 16—Rockwell Hardness Survey of Steel X30864 Hardened as Shown. P-F Type 1, See Table VIII.

steel and the uniformity in the hardened zone as shown in Fig. 19 and Table IX, and this feature is being subjected to further investigation.

Modified P-F Test

Grossmann* found that "each different lot of steel had its char-

*M. A. Grossmann, "On Grain Size and Grain Growth," Transactions, American Society for Steel Treating, Vol. XXI, Dec. 1933, No. 12.

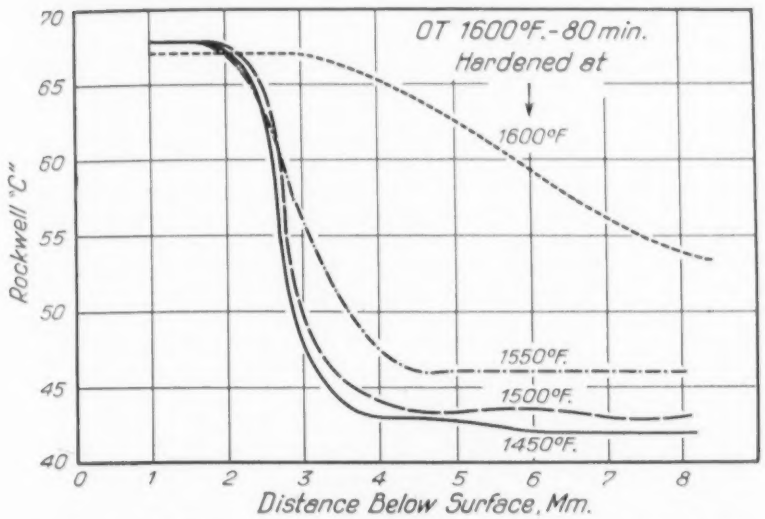


Fig. 17—Rockwell Hardness Survey of Steel X29983 Hardened as Shown. P-F Type 3, See Table VIII.

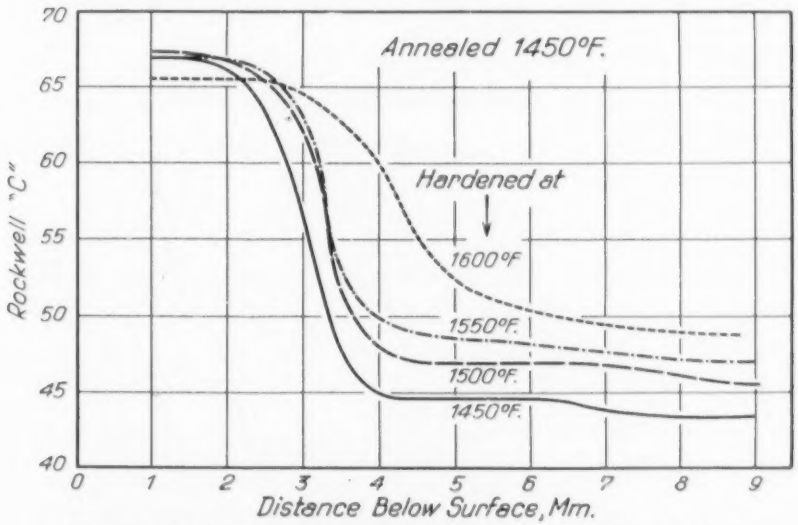


Fig. 18—Rockwell Hardness Survey of Steel X29983 Hardened as Shown. P-F Type 3, See Table VIII.

acteristic coarsening temperature." "If the carburizing temperature is conducted at any temperature below the coarsening temperature . . . . . , then the austenite grain-size will remain unchanged, precisely what it was when the steel first transformed to austenite upon heating." It is evident that the McQuaid-Ehn grain-size test requires a standard carburizing temperature for uniform results. This has been set at 1700 degrees Fahr., which is higher than usually used in commercial carburizing.

It was thought that the P-F test could be modified for carburiz-

ing and other steels so that the results would indicate the hardness, core hardness and fracture which would be obtained in commercial utilization of the steel. A similar test was made to the P-F test previously described, excepting that the samples were carburized

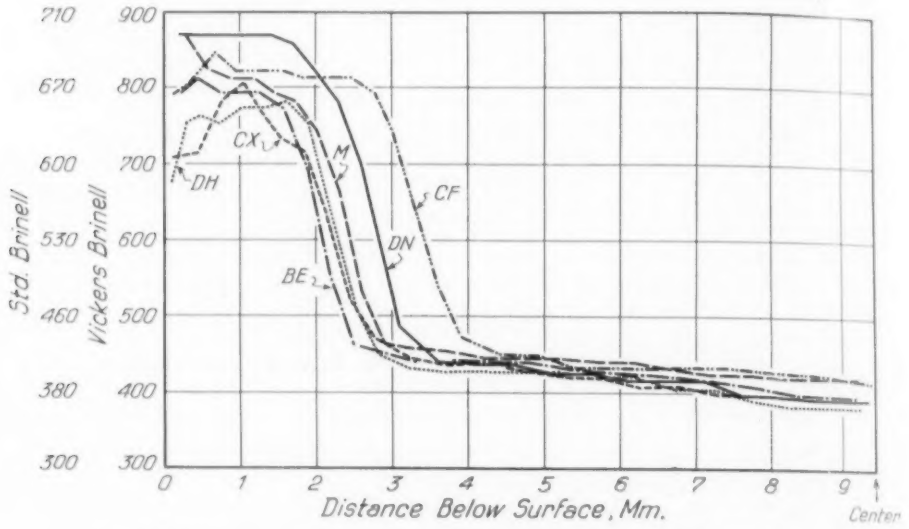


Fig. 19—Vickers Hardness Survey of ¾-Inch Round P-F Test Pieces. See Table IX.

for 16 hours after heated through at 1640 degrees Fahr. (a temperature commonly used in our factory), cooled in the pots and hardened according to the standard P-F method.

Table X
Comparison of A.S.T.M. Grain-Size Classification with Modified P-F Test of S.A.E. 1020 Simple Carbon Carburizing Steel

Manufacturer	A.S.T.M. Grain Size	Case Fracture Grain-Size —Degrees Fahr.— 1450	1500	1550	1600	Core Brinell Hardness —Degrees Fahr.— 1450	1500	1550	1600
C	3	6¾	6¼	6	4¾	192	192	207	207
A	3	6¾	6¼	5¾	4½	207	207	229	229
A	4	7	6½	6¼	5½	212	207	235	229
A	8	7¾	7½	7¼	7¼	207	207	229	229
B	8	7½	7¼	7	6¾	201	207	235	235

Fracture grain-size tests were made of the case and Brinell hardness tests made in the center of the core.

It will be noted from Tables X and XI that the McQuaid-Ehn test bears a close relation to the fracture grain-size obtained on the modified P-F test. This indicates that observation of the fracture

Table XI

Comparison of A.S.T.M. Grain-Size Classification with Modified P-F Test of S.A.E. 4615 Nickel-Molybdenum Carburizing Steel

Manufacturer		A.S.T.M. Grain Size	Case Fracture Grain-Size —Degrees Fahr.— 1450	1500	1550	1600	Core Brinell Hardness —Degrees Fahr.— 1450	1500	1550	1600
A	.17 C .52 Mn	8	8¾	8¼	8¼	7¾	363	363	388	388
B	.16 C .51 Mn	8	8¾	8¾	8¾	8¼	285	302	302	302
C	.16 C .51 Mn	5 & 8	8¾	8	8	7	255	269	277	277
E	.18 C .54 Mn	7	7	6¾	6	5¾	363	375	375	388
D	.17 C .61 Mn	6	7	6¾	5¾	5	302	311	311	321
E	.16 C .40 Mn	3	6	5½	4¾	4	285	321	321	321

grain-size of carburized articles during commercial hardening offers a quick, ready and accurate method of determining "grain-size" without reference to the microscope.

Acknowledgments

Due acknowledgment is made to the metallurgical staffs of Ingersoll-Rand Co., Halcomb Steel Co., Bethlehem Steel Co., Carpenter Steel Co., and SKF Steels, Inc., for assistance in this work.

DISCUSSION

Written Discussion: By A. E. Focke, research metallurgist, Diamond Chain Co., Indianapolis.

Three criteria which may be used to judge the merits of any test are: 1. The importance of it; 2. Its accuracy and the ease with which it may be reproduced; 3. The extent to which the purpose of it is realized.

Any test which has for its purpose an attempt to define or classify that quality in steel which is variously described as "body" "timbre" and "personality," is vitally important to a user or manufacturer of heat treated steel articles.

Mr. Shepherd has devised a relatively simple method of studying, classifying, and correlating two of the important characteristics of a heat of steel. It is obvious that he has carefully investigated the limitations of his method and established a definite procedure which must be followed. However, it appears significant that the author admits that the determination of the fracture

part of his P-F test by visual comparison with a set of standards requires a high degree of personal skill.

The foregoing, together with the fact that I would personally react more favorably to some test method, which would permit a grain-size analysis, without the use of an arbitrary set of standards, are my principal criticisms of the author's test method. I was somewhat disappointed that Mr. Shepherd did not illustrate the value of his test by more definite references to specific problems for which he has found it to be useful.

I am particularly concerned with the "personality" of the carburizing grades of steel and I was therefore, greatly interested in the author's modification of his P-F test to meet this requirement.

An analysis of the results recorded in Table XI reveals some rather interesting information.

There appears to be no simple relation between A.S.T.M. grain-size or for that matter Case Fracture grain-size and the core strength. Neither is there any agreement between the amounts of the common elements as indicated by the regular chemical analyses and the core hardness developed after heat treatment. Admittedly, the factors of normality and abnormality must play important parts in controlling this core reaction but measuring the relative degrees of normality of alloy steels of this class particularly in the fine grain-sizes is not simple. We have been forced to recognize these facts and have had to add core strength after hardening requirements to our specifications for certain critical parts.

It can be seen that there is much greater difference in the core strength of different steels after the same treatment than there is between the core strength of the same steel after quenching from different reheating temperatures. (Maximum difference between core Brinell of the same steel after quenching from 1450 and 1600 degrees Fahr. was 36—Minimum 17—Average 24). (Maximum difference between core Brinell of different heats after quenching from 1600 degrees Fahr. was 111). This result confirms our experience in attempting to salvage heats which showed insufficient core strengths by simple variation in the reheating temperature.

I question, therefore, that the amount of additional information obtained from testing this type of steel at the four temperatures over that obtained from using one temperature would justify the additional labor.

But, it occurs to me to wonder if the author has not, by limiting his "penetration factor" in his modified P-F test to a simple record of the core Brinell hardness after reheating, overlooked the important variables of depth of case penetration and gradation of case hardness which, of course, are very important factors in determining the value of a part for service.

The amplification of the author's modified P-F test to include these variables appears to me to have interesting possible applications to my own problems and I therefore, wish to express my appreciation to the author for his paper.

Written Discussion: By H. G. Keshian, metallurgist, Chase Companies, Waterbury, Conn.

The author states at the bottom of page 989 that hardenability is a heat characteristic. Amplifying this point further in the second sentence of the first

paragraph on page 890 he says, "This indicates that P-F characteristic is due to the melting reactions and not to the ingot size, temperature of pouring, etc." This statement of the author would mean that the different bars of steel from a given heat will have the same hardenability characteristic. This, however, does not seem to be the case, because at times, from the same heat of steel, not only the different bars but also the different ends of the same bar may show considerable difference in hardenability, due mostly to the degree of segregation in the different ingots, and in the different ends of the same ingot. If this is so, contrary to the statement of the author, the size of the ingot, that is, the rate of solidification and the temperature of pouring (because these influence the degree of segregation) must all have an effect on the hardenability of the material. Consequently, because a part of a heat of steel shows a satisfactory hardenability, that does not mean that the rest of the heat has the same hardening characteristic.

Furthermore, as the author very correctly states on page 890 the necessity of testing the first and last ingots from the same heat also shows that the fact that various bars of steel are made from a given ingot is no indication of uniformity, except, of course, for chemical composition. So, when considering the effect of a heat of steel from the standpoint of uniformity, some other tests such as hardenability, segregation, or macro-structure have to be applied to each bar to verify the uniformity of the various characteristics. This still leaves the steel mill practice as the best guarantee for the uniformity of hardenability, macro-structure, etc.

Written Discussion: By F. H. Allison, Jr., research metallurgist, Crucible Steel Company of America.

From the standpoint of tool steel, and particularly of carbon tool steel, Mr. Shepherd's paper is of extraordinary importance. It has long been recognized that fundamental differences exist in carbon tool steel heats melted under different practices. Mr. Metcalfe, some thirty or forty years ago, on the same site at which the writer works at present, devised the Metcalfe or fracture test familiar to all tool steel men, in order to distinguish between the various melts of carbon tool steel. All accredited tool steel manufacturers have long had their individual test methods to determine and classify the body or quality of tool steel; but it has remained for Mr. Shepherd to develop and publish a method which bids fair to be accepted as a standard for the industry.

For the reason that the author's method is likely to become a standard procedure for an entire group of steel suppliers and consumers, it merits the most careful consideration and criticism in order that it may be applied as widely and fully as possible. The following remarks are offered with this purpose in view.

To be able to refer to a standard set of fractures in which a No. 9 means the same in every shop, mill and laboratory, will come as a relief to metallurgists and tool steel men who have long been troubled by the vagaries of "fine," "medium" and "coarse." Mr. Shepherd has done a service, indeed, if the graded fractures are established as a standard of reference, and the writer heartily endorses their use.

Concerning the designation of fracture, it is felt that the term "grain-size" should be avoided, since this term already has a significance in metal-

lography which easily leads to confusion if adopted also in reference to fracture. The term "fracture grain-size" is little better since in discussion it is extremely easy to omit the word "fracture." It is suggested that "fracture appearance," "fracture texture" or some better term be used for the sake of clarity in these references.

In regard to penetration values, the author reports them in terms of 1/64th inch or half of 1/64th inch. There are advantages in reporting these measurements in 1/100th inch and these should be carefully considered before the method is too rigidly standardized. The accuracy of measurement is about the same in using either 1/128th inch or 1/100th inch. By reporting the true hundredths of an inch penetration the use of fractions is avoided, the magnitude is apparent in the actual figures, and the decimals readily distinguish the penetration from the fracture values in the full P-F report. Thus a steel reported by the present system as:

5½, 6½, 8, 14 : 8¾, 8½, 8¼, 7¾

would become:

.09, .10, .13, .22 : 8¾, 8½, 8¼, 7¾.

The gage shown in Fig. 4 of the author's paper could be as easily calibrated in hundredths as in sixty-fourths of an inch.

Further work in regard to penetration measurement appears to be necessary in that the transition zone between the hardened case and core may, as the author points out, be several hundredths of an inch in width, so that the exact boundary of the hardened zone is not always easily determined.

The splendid conception of grouping the change in penetration with the change in fracture appearance as the hardening temperature is increased, affords a very full picture of the behavior of the steel. The further classification of steels into types such as 1, 2/1, 3, etc., may, however, be satisfactory for individual application and for simplification; but, for general application it appears with the present knowledge to be somewhat premature and may also be misleading. In the first place, it is felt that wider experience over a greater number of tool steel applications should be obtained before fixing the arbitrary limiting values assigned to the several types. Secondly, a single type as it now stands may embrace steels of widely different character. For example, the P-F characteristics of the following two steels would both be classified as type 1, yet the first is shallow hardening with fine fracture, and the second is relatively deep hardening with coarse fracture.

	Penetration		Fracture
First	.08, .08, .10, .11	:	9, 9, 9, 8½
Second	.14, .14, .16, .17	:	6, 6, 6, 6½

The author calls attention to the dumb-bell type of core in Fig. 8 found on hardening certain disks. Fig. 9 indicates that this condition is more likely to be characteristic of the portion of the ingot from which the sample was taken, rather than characteristic of the entire heat.

The paper contains data which are decidedly pertinent to the general subject of hardenability. The supposition has been advanced by others that the depth of penetration depends upon austenitic grain-size and that the greater the grain-size, the deeper will be the hardened case. Similarly, it has been

assumed that the fracture appearance is due to the austenitic grain-size—the coarser fracture being occasioned by larger original austenitic grains. However, from the data of the present paper, it is possible to find a shallow hardening steel possessing a coarser fracture than a deep hardening steel or vice versa. In fact, the depth of hardening and the fracture appearance seem to be entirely independent of ane another as can be seen from the following data selected from the paper:

Heat	Penetration		Fracture			
H5194	.09, .09, .11, .13	:	9,	9,	9,	8½
5760	.08, .09, .10, .12	:	8½,	7¾,	5,	4½
2 H	.22, .22, .22, .23	:	8,	8,	7¾,	7¾
30420	.14, .17, .38, .38	:	8¼,	7¼,	6¼,	5½

Evidently, the growth of the austenite grains cannot be the cause of the difference both in hardenability and in fracture, or there would be more correlation between the two than above indicated.

In concluding, the author has made a valuable contribution to the metallurgy of tool steel, both in providing an excellent test method and in emphasizing that there exist a number of types of carbon tool steel, independent of composition, that are available to the tool steel user to suit his various requirements.

Written Discussion: By R. Schempp, metallurgist, Halcomb Steel Co., Syracuse, N. Y.

In his present publication, Mr. Shepherd has given both the steel maker and the consumer another valuable and very practical means for the determination of the inherent hardening, grain-size and temperature sensitivity characteristics of steels, especially tool steels. Unlike the rather unfortunate choosing of "normal" and "abnormal" as a means of conveying the structural grain-size characteristics of the hypereutectoid zone as developed after the standard McQuaid-Ehn carburizing treatment and observed under the microscope, Mr. Shepherd's terminology and method of designation as well as his test procedure, distinguish themselves by their extreme clearness, simplicity and accuracy and lend themselves to a quantitative measure of the properties involved.

Speaking of the inherent hardenability and P-F characteristics of tool steels of identical composition, it may be well to point out that while these properties are essentially the result of certain melting practices, these so-called inherent characteristics of hardenability and martensitic penetration upon quenching, with but few exceptions (that is, on very stable types) are materially influenced by

1. Prior structure.
2. Critical rate of cooling during the quenching operation.
3. Mass action.

The apparent contradictory influence of prior structure upon the hardening properties of tool steels when using a low (1450 degrees Fahr. or high 1600 degrees Fahr.) hardening temperature, may be explained by the predominating influence of carbide solubility as brought about by prior structure

in one case and by the stronger influence of the austenitic grain-size prior to quenching in the other.

The early results of test work which is being carried on in our laboratory indicates a very definite relationship between the austenite grain characteristics and the hardenability and P-F values as determined by the Shepherd methods. On tool steels, Mr. Shepherd's testing procedure will allow us to definitely classify heats as to their specific types. Our practice has been to differentiate between the following four types:

1. Material on which the penetration of hardened case as well as the grain-size is stable, and will not increase with increasing temperature.
2. Material on which the penetration of hardened case will increase in depth with increasing temperature, the grain-size of which, however, will not coarsen.
3. Material which will not increase in penetration of hardened case with increased quenching temperature, but the grain-size of which will coarsen.
4. Material on which both the penetration of hardened case and grain-size will increase with increasing quenching temperature.

Neither by this means nor by any other means of testing which has been developed in the past has it been possible to determine the physical property characteristics as ductility, toughness, etc., of this class of material in the hardened condition in such a manner as would enable us to differentiate between heats of similar composition in regard to their toughness in the hardened state, and to predict certain service results for various applications.

In his efforts to develop a more satisfactory heat treatment for our high speed steel grades, Mr. Kingsbury, a member of our metallurgical staff, applied as early as 1911 the static load fiber stress test as later described by E. G. Herbert under "The Influence of Heat on Hardened Tool Steels" in the *Journal of the Iron and Steel Institute,* Vol. 85, #1, of 1912, p. 358 to 378, to determine the toughness characteristics of high speed steel. Herbert's determinations were primarily confined to maximum fiber strength alone. Strength, however, is not necessarily a measure of toughness. This comparatively simple means of obtaining the maximum fiber stress values on hardened material has now been modified and developed into a testing procedure which lends itself remarkably well for an actual determination of the inherent toughness values of hardened high speed and tool steels.

This essential "it" of tool steels, may it be termed toughness, body or any other name, is an inherent heat characteristic which all of us have learned to know and to value, but for which we have never found a means of expression and determination. Our new method tests the outermost fibers of the specimen under conditions of comparatively heavy and rapid loading and while our test is admittedly a static test as against the well recognized fact that tools in service fail through dynamic stresses, we believe that our method very closely approximates actual service conditions.

In our opinion toughness must be considered as a combination of strength and ductility and, as we will show in the following results which have been

obtained on several heats of straight carbon, general purpose tool steel, we have expressed this inherent toughness in the form of a coefficient of toughness value.

To eliminate the influence of prior structure, we also resort, as does Mr. Shepherd, to a preliminary oil treatment before hardening our specimens which are usually ⅜ inch round and approximately 6 inches long. In the case cited the samples were hardened from 1550 degrees Fahr. and tempered at 415 degrees Fahr. for one-half hour.

In the following table we give the composition of the material used, its characteristics as obtained by the Shepherd hardenability and P-F method as well as the normality and grain-size rating determined according to the McQuaid-Ehn test.

Composition:

Heat No.	C	Mn	P	S	Si
H-4705	1.08	0.23	0.017	0.018	0.29
W-5294	1.05	0.21	0.011	0.015	0.24

Results of regular testing procedure:

Heat No.	Hardenability Rating (slab test)	P Value	F Value	Type	McQuaid-Ehn Test (A.S.T.M. Chart)
H-4705	10	6, 7, 8, 12	8¾, 8½, 8¼, 8¼	2/1	Abnormal 6/7 Grain
W-5294	10	7, 8, 10, 24	8¾, 8¼, 7, 6¼	3	Abnormal 5/6 Grain

Kingsbury Toughness Test:

Heat	Fiber Strength	Ultimate Deflection	Coefficient of Toughness	Rockwell Hardness
H-4705	222,500	0.094	21,150	62.5
W-5294	181,000	0.0674	12,200	62

We may add that Heat H-4705 was reported as being very satisfactory and tough bodied in service; while the material of Heat W-5294 proved brittle and unsatisfactory.

This new test has given us a number of very convincing confirmations from its practical application. We have mentioned this new test because it seems to us that it supplements Mr. Shepherd's hardenability and P-F testing procedure.

Complete details of this toughness test for hardened tool steel materials will be published in the near future.

Written Discussion: By L. D. Hawkridge, Hawkridge Brothers Co., Boston, Mass.

Again we are indebted to Mr. Shepherd for a most interesting and important contribution, disclosing a simple and thoroughly practical method of classifying tool steel in accordance with the hardening characteristics—a method, which, in the writer's opinion, will take its place beside the McQuaid-Ehn test as one of the indispensable tools of the industry.

While the manufacturer or distributor of tool steel may deplore the advent of another test containing a large measure of the "personal equation" (which there undoubtedly is in the estimating of the F values), he must admit that by no other means known at present could so much, or so exact information, be packed into so little space as by the series of eight numerals proposed by Mr. Shepherd.

The latter himself wisely stresses the difficulty of rating fractures correctly, and advocates special training of those who will act as observers. This is important, as the writer can testify from the few attempts that he has made to score over 80 per cent in the fracture rating test. Mr. Shepherd might well have stressed also the fact that luster often tends to obscure grain-size. It is a common error of beginners to rate a fiery fracture too far toward the coarse end of the scale. But in spite of the difficulties that one can foresee in the application of this test, it will be by no means more difficult to obtain agreement between buyer and seller than in other tests now widely used—for instance, the hot acid etch test.

Of particular interest to the writer is the comparison between the P-F values of the 8-ton basic electric heat and the 100-ton basic open-hearth heat, as given in Tables VI and VII. How does Mr. Shepherd account for the apparent instability both of the grain-size and penetration of the specimens from the open-hearth heat which were hardened from 1550 degrees Fahr.? Here is a variation of 50 per cent in the penetration of hardness in specimen D 7 as against D 4, whereas the maximum variation between ingots of the electric heat at the same temperature is less than 10 per cent. Why should the grain-size vary so widely between specimens of the open-hearth heat quenched from 1550 degrees Fahr. when those quenched from 1600 degrees Fahr. are in much closer accord?

One more question—Do not the results set forth in Table VIII indicate that, when a test is being made of Type 3 steel, it is most important to maintain the specimens at 1600 in the normalizing for the full 80 minutes recommended earlier in the paper? If this soaking time is cut down, will not results be questionable because of non-homogeneity of the pre-hardening structure? Incidentally, why should the relatively coarse particles of carbide in pearlite (presumably spheroidized) inhibit grain growth at high temperatures more than the very finely divided carbides in sorbite?

While particularly applicable to tool steel, Mr. Shepherd's method, as modified for carburizing steels, will be decidedly useful, and there would seem to be no reason why a similar modification could not be worked out for steel of intermediate carbon ranges both with and without alloys. With such broad fields for further investigation opening out, we should be able to look forward to many more illuminating papers from Mr. Shepherd on the same general theme.

Written Discussion: By G. V. Luerssen and O. V. Greene, Metallurgical Department, The Carpenter Steel Co., Reading, Pa.

Probably no physical characteristic of steel is more actively discussed and less accurately reported than the fracture test. We are certainly indebted to Mr. Shepherd for giving us standards for the evaluation of a very useful test.

The phenomenon observed by Mr. Shepherd on specimens pre-hardened by oil treating from 1600 degrees Fahr. is particularly interesting. Table VIII, which reports P-F characteristics on types 1 and 3, shows that heating specimens to 1600 degrees Fahr. and oil quenching, followed by brine quenching, produced a shallower penetration than was obtained in specimens brine-

quenched from the annealed condition. This was true only at temperatures of 1450 and 1500 degrees Fahr. After this temperature range was passed, and higher temperatures such as 1550 and 1600 degrees Fahr. were employed, the opposite was the case, and the preliminary oil treatment tended to cause a deeper penetration and a coarser fracture. These characteristics were particularly noticeable in the heat of type 3.

It is quite possible that these results may be explained by considering the structure of these specimens before the final hardening. A steel of type 1 after oil treating probably retains some of the eutectoid carbide in solution, with the balance submicroscopic or nearly so in size. This treatment also disperses and reduces the size of the hypereutectoid carbide. Consequently, when a specimen of type 1, oil treated from 1600 degrees Fahr., is heated for final brine quenching from 1450 degrees Fahr., for instance, the first austenite will form where the greatest amount of carbides are in solution or where the spheroids are finest. These early crystals of austenite must, therefore, be exceedingly small and very numerous. The balance of the ferrite and some of the undissolved carbide will then be absorbed as the austenite grows. However, at this temperature there will remain a comparatively large number of widely distributed carbides. These well dispersed particles not only impede the grain growth, but act as nuclei which accelerate the transformation and, consequently, produce a decrease in the hardenability.

When an annealed specimen of a steel of type 1 is heated for hardening at 1450 degrees Fahr., a similar reaction takes place, except that the formation of the austenite is considerably more difficult, and the original islands are fewer. The undissolved carbides impede the grain growth as before, but are not sufficiently dispersed to offer enough nuclei to accelerate the transformation. Consequently a slightly deeper hardenability is exhibited.

As a matter of fact the rate of transformation of oil-treated specimens on heating is so increased, or the actual temperature of transformation lowered, as reported by Mr. Shepherd, that it can be easily observed in an ordinary heating furnace. If an oil-treated and an annealed specimen be placed in a furnace at 1450 degrees Fahr., and observations made during the heating, it will be noticed that the oil-treated specimen will be through its critical and nearly up to furnace temperature, while the annealed specimen will still appear quite black.

The structural changes occurring in an oil-treated specimen of type 1 steel during heating for final brine quenching at 1600 degrees Fahr. are similar to those that took place when an oil-treated specimen was heated to 1450 degrees Fahr. However, as the larger austenite islands grow, they absorb more and more of the smaller and less favorably situated islands, until all the grains have reached a fairly uniform size. The eutectoid and hypereutectoid carbides in this specimen are so small and so well dispersed that they are readily and completely soluble at 1600 degrees Fahr. Consequently with no particles to arrest the grain growth or act as points of crystallization on quenching, the specimen exhibits a deeper hardenability and a coarser fracture than the one brine-quenched from 1450 degrees Fahr. in the oil-treated condition.

The annealed specimen of type 1, on the other hand, when heated for brine-quenching at 1600 degrees Fahr. does not dissolve all of the hypereutectoid carbide. The hypereutectoid carbides in this specimen are comparatively large and difficultly soluble. They impede the grain growth on heating and offer points of crystallization on quenching. Consequently the presence of the carbides in the annealed specimen brine-quenched from 1600 degrees Fahr. produces a finer grain and a shallower hardenability than can be obtained in the oil-treated specimen brine-quenched from the same temperature.

The dissimilarity between the P-F characteristics of the oil-treated and the annealed specimens of a steel of type 3 is due to the same causes and are differences of degree and not of kind.

Type 3 steels exhibit a much deeper hardness penetration and a much coarser fracture than type 1 steels, because of inherent differences. The repression of grain growth in type 1 steel, particularly in the high temperature range, is due to the presence of widely dispersed submicroscopic nonmetallic particles. Such particles do not dissolve or coalesce at the higher treating temperatures as do carbides, and, consequently, they continue to obstruct the grain growth. These particles also offer nuclei for the transformation of austenite on quenching. Thus these minute nonmetallics impede the grain growth and accelerate the austenite transformation to such an extent that a comparatively fine fracture and shallow penetration is obtained when a steel of type 1 is quenched from the higher range. A steel of type 3, which is dependent upon the presence of carbides alone for the impedance to grain growth and for the acceleration of the austenite transformation, shows a coarser fracture and a deeper hardness penetration when quenched from the higher treating temperatures.

Written Discussion: By A. H. d'Arcambal, consulting metallurgist, Pratt and Whitney Co., Hartford, Conn.

Our knowledge of tool steels has been greatly increased during the past few years as a result of the published research work of such able investigators as Shepherd, Gill, Emmons, Luerssen, Wills and others. Mr. Shepherd's paper on "The P-F Characteristic of Steel," just presented, should prove of interest and value both to manufacturers and users of tool steels. He again definitely shows that steels of the same chemical composition when subjected to similar hardening treatments react differently, certain steels hardening much deeper than others, a coarser grained fracture also being noted on some of the types tested. We would be interested in learning the composition and the treatment given each of the ten specimens shown in Fig. 1. Another set of masters made from a different heat of steel of the same composition, especially if produced by a different mill, undoubtedly would be quite different in fracture grain-size for corresponding samples. Is it not quite difficult, moreover, for two observers to check the "F" factor within half a fracture number?

We would be interested in knowing if all of the carbon tool steel purchased by Mr. Shepherd's company is subjected to this so-called "P-F" test. Should this be the case, has not the author found that tool steel of the same chemical composition from the same mill very rarely does not pass this "P-F" test from heat to heat. We have found that tool steel, heat after heat, from

our principal source of supply shows a satisfactory fracture and hardness depth when quenched from elevated temperatures, but tool steel of the desired composition from other mills will not always react so favorably to these valuable tests first suggested by the author some years ago.

We noted with considerable interest Mr. Shepherd's comments relative to a comparison between electric furnace and lead pot method of heating. It has been our experience that samples heated in the lead bath come to temperature very rapidly as compared with similar sized parts heated in the electric oven type furnace. For example, tool steel specimens ¾ inch diameter by 6 inches long will harden satisfactorily after being heated in the lead bath at 1460 degrees Fahr. for only one minute. Similar sized specimens from the same bar of steel when heated in the electric furnace required 15 minutes to reach the furnace temperature of 1450 degrees Fahr. The specimens heated in the lead bath as well as the electric oven furnace treated samples were brine-quenched, after the one-minute and fifteen-minute times, respectively, resulting in a Rockwell hardness in each case of C-66-67. The samples treated in the electric furnace showed a slightly greater hardening depth than the samples given one minute in the electrically heated lead bath furnace. Specimens given longer periods of time in the lead bath showed the same hardening depth as the specimens heated for one minute.

It will be interesting to note the reaction of the tool steel manufacturers and users to the author's suggestion relative to purchasing tool steels on the penetration-fracture basis as outlined in this interesting paper by Mr. Shepherd.

Written Discussion: By S. C. Spalding, metallurgist, American Brass Co., Waterbury, Conn.

The author is to be congratulated on originating a practical method of classification of carbon tool steels. Although tool steel makers and users have talked of inherent quality, body, timbre, etc., it was apparently left to Mr. Shepherd to give us a practical and simple method and set of standards for their determination. The general adoption of this method of classification should help materially to bring up the average quality of our carbon tool steels. A consumer today finds only to his sorrow that steels of the same analysis and supposed quality purchased from different sources vary widely in physical characteristics. When he and the mill metallurgist get together there is seldom an agreement, even in terms to describe the fracture characteristics.

Here we have a set of standards set up for us. With one of these boxes at hand the matters of difference could readily be settled. We hope the steel producers will be sufficiently impressed by this paper to equip themselves with the standard fracture box and grade their steels accordingly.

It would seem to us a great step forward to be able to purchase steel to a certain range of P-F characteristics, rather than to a brand name. I would like to recommend that our society of itself, or in conjunction with the A.S.T.M., make a study of this method, with the idea of working out acceptable standards for inclusion in specifications for tool steels.

Written Discussion: By A. C. Jones, research engineer, Lebanon Steel Foundry, Lebanon, Pa.

Mr. Shepherd's contribution to that phase of metallurgical science dealing

with grain-size control and hardenability is not only interesting but extremely valuable. His whole scheme, cleverly worked out, must appeal not only to the research engineer, but also to the control metallurgist, and anyone interested in getting the most out of his steel product.

To get the evidence that Mr. Shepherd has presented, was probably less difficult than to explain the causes of the uniformities shown. Undoubtedly the method and degree of deoxidation and degasification in the melting history of the steel are very important factors. His scheme applied to the observations in the steel plant should help establish the important factors affecting the grain-size and hardenability.

I would like to ask Mr. Shepherd if any relation has been observed by him between "inclusion count" and grain-size or hardenability. Also, was there any notable difference in effect between steels with large inclusions and those with small inclusions? Another question—has he any data to indicate that the fine-grained carbon steel preserves most of its impact resistance at sub-zero temperatures?

Written Discussion: By Henry Wysor, metallurgical engineer, Bethlehem Steel Co., Bethlehem, Pa.

We are greatly indebted to the author who has given us a new definition for steel quality, based on obvious phenomena of hardening. The experimental work by which the data was obtained is admirable in both the extent and the care to which it was prosecuted. While the primary object was to test steel quality in the improvement of heavy duty tools, the usefulness of the "P-F" test in other lines of developments and research is foreshadowed.

It should prove very interesting to compare the "P-F" characteristics of carbon and special steels with fatigue values. Take for example spring steels, especially carbon and low alloy spring steels. There appears to be a decided connection between penetration characteristics in quenching and endurance of springs in vibratory tests provided the section is heavy enough to develop a core fracture. That is simply an observation which I have made on both spiral and leaf type springs, but I propose to apply Mr. Shepherd's method of testing for a more definite proof.

Written Discussion: By Thomas G. Foulkes, metallurgical supervisor, Bethlehem Steel Co., Bethlehem, Pa.

The test, so fully developed by Mr. Shepherd for determining the so-called P-F characteristic of steel, and so clearly presented in his paper on this subject, will have an especial appeal to the heat treatment shop man. Here is a test that he can understand and appreciate. This test eliminates, for him, the mystery of a laboratory report couched in scientific terms, as it gives him an instrument for visual classification.

The testing procedure, while exacting, is not difficult. The adoption of this test will give impetus to improvements in the application of steel for treated parts, making possible the consistent use of a selected type of steel based on P-F values, determined from comparison of parts having given satisfactory as well as unsatisfactory service.

It is realized that all types will be in demand, the type desired depending on the particular application involved. The steel industry stands ready to

co-operate in the development of the proper type for particular applications.

The writer is acquainted with the consistent production of two of these types, listed by Mr. Shepherd, over a period of several years.

The Shepherd P-F characteristic test is a distinct contribution to the advancement of steel industry technique, giving a ready and proved method for classification and selection of steels.

Oral Discussion

GILBERT E. DOAN[1]: Is there any correlation between the P-F characteristic and the gas content of the steel in question? For example, is there any relationship between the oxygen and nitrogen content of the steel and its P-F characteristic? Furthermore, have you found that there is any relationship between the above mentioned characteristic and the grain-size of steel as measured in any of the conventional ways?

DR. J. A. MATHEWS[2]: On the bottom of page 983, Fig. 2, where he refers to normalizing as samples heated at 1600 degrees Fahr. and quenched, I think that should be called pre-quenched. The society of which Mr. Shepherd is now president, and two other societies, have a committee on definitions, but I think we should stick by our old definitions.

Author's Closure

In reply to Mr. Focke's discussion may we say that our particular applications of the P-F characteristic test are for applications in a field which is very highly specialized and would not be of direct value to any one except those few persons engaged in that particular industry.

The applications of the test must, therefore, be worked out for each individual's particular requirements.

The use of four temperatures serves to accentuate any differences which may exist and the test in reality grew from the inability of one temperature to give sufficient information. We believe that the test has not, as Mr. Focke states, been developed beyond its initial stages. The width of its application can only be determined by further work.

As Mr. Keshian indicates it would be quite easy to find differences in hardenability and P-F characteristic throughout a heat. We must, however, consider that the particular heats under observation represent carefully refined, melted, poured and rolled steel in which good practice has been used to such an extent as to control all of these factors within the limits which are possible at the present time. The acceptable bars from heats of steel of this type would show remarkable uniformity with regard to hardenability and P-F characteristic If any marked non-uniformity would be shown by these tests it would indicate that a change in the source of supply should deserve careful consideration.

As Mr. Keshian states, the steel mill practice is the best guarantee of the uniformity of hardenability, microstructure, etc., but it is very necessary

---

[1]Instructor in metallurgy, Lehigh University, Bethlehem, Pa.

[2]Vice-President, Crucible Company of America, New York City.

to continually check the results of this practice by incoming material inspection tests.

Mr. Keshian has touched on one point which is probably not emphasized or explained in detail in the original paper. This new test method is primarily used to determine the difference between different heats of steel of approximately the same chemical composition with the idea of finally correlating service results with a particular P-F characteristic. It is essential, therefore, to use metal representative of the heat of material to be tested and for this reason we forge the stock to approximately ⅞-inch round. Tests made on an ingot from the outside and interior of an ingot, from the outside and interior of a billet, from the outside and interior of a bar, would very probably show a considerable difference in P-F characteristic, diminishing as the steel was reduced in size. For this reason any tests made in a manner different than that described on page 983 should be accompanied by specific detailed data.

In reply to Mr. Allison's discussion may we say that fracture grain-size is used to designate the texture of the fracture of the P-F test pieces because it is closely allied to the actual martensitic grain-size of the test piece. There is no more excuse for omitting the word fracture in referring to the fracture grain-size, than there is for omitting the word austenitic in speaking of austenitic grain-size, or in omitting A.S.T.M. in referring to A.S.T.M. grain-size classifications, etc. It does not appeal to us to refer to the fractures as fracture appearance #8 or fracture test #6. The grain-size is an actual grain-size, although the scale is empirical.

The penetration values are reported in $\frac{1}{64}$ of an inch because the original Shepherd slab test known to the steel trade for several years was based on $\frac{1}{32}$-inch variations in thickness. Since the penetration is measured on one side only, it was only natural that we should continue our designation of hardenability in the same terms in which it was originally conceived, namely, ½ of $\frac{1}{32}$ or $\frac{1}{64}$.

The suggestion of one hundredths of an inch should be an optional choice with the user of the test. We have accepted Mr. Allison's suggestion and are grading our gage in both scales.

It is true that a single type may embrace steel of widely different analysis but the P-F characteristic type may be the same and it is felt that the classification of the P-F values into types is justified through its clear indication as to whether the steel is stable or unstable with regard to either penetration or fracture grain-size, or both. For example, we may have a 0.70 per cent carbon tool steel which is the same type as a 1.05 carbon tool steel, but its P-F characteristic will be different. It has been found that certain of the inherent properties of these steels are similar in regard to toughness, etc., but, of course, they may be of entirely different hardenability and P-F characteristic throughout.

In reply to Mr. R. Schempp's discussion we look forward with interest to complete details of the toughness test for hardened tool steel material, particularly as these materials will be identified by their particular P-F characteristic.

After this information has been published it is entirely probable that a

great many of our tool steel users will see much more clearly the necessity for obtaining data an tool behavior of steels which have been identified by their P-F characteristic and not by the indefinite approximate chemical composition. After such correlation it is entirely reasonable to expect that new specifications for tool steel will have as their essential requirement a test similar to that described in the paper.

In reply to Mr. Hawkridge's discussion may we say that unless the observers have had experience in judging tool steel fractures it, of course, is too much to expect to have a novice read fractures with the skill required by the fracture portion of this test. It has been necessary to apply special training of probably a much more intensive character to our students in metallography. Skilled tool steel men usually have very little trouble in passing 80 per cent in the fracture rating test, provided proper light conditions have been available.

We believe luster is the result of grain-size. The shininess of a surface depends upon its ability to reflect light in unbroken wave fronts. Fine-grained specimens have a tendency to be more lustrous than coarse-grained specimens.

The specimens from the open-hearth heat shown in Tables VI and VII were hardened through. 1550 degrees Fahr. was a border line where a more or less degree of variation in hardness might be expected. The electric furnace heat shown in Fig. 13 is of a different type and hardens much less deeply at 1550 and 1600 degrees Fahr. The less the depth of hardening the more nearly the specimens may be expected to be in accord, all other factors being equal. We have found it necessary to maintain the specimens at the 1600 degrees Fahr. normalizing temperature for at least 40 minutes after heated through. We agree with Mr. Hawkridge that a modification of the P-F test would be extremely valuable for steel of intermediate carbon ranges, both with and without alloys. We have offered the test with the hopes that some of our more able investigators will carry this work on.

In reply to Messrs. Luerssen and Greene's discussion there is a question in our mind whether Type 1 material produces a less soluble carbide or whether the solute is less active. Regardless of this, we are convinced that Type 1 material is not "carbide sensitive" because the presence of other penetration propagation inhibitors, such as submicroscopic oxides, etc., is more pronounced than in the other P-F types of steel of the same composition. Type 1 is not annealing sensitive and, therefore, the question of carbide size does not enter the picture.

The balance of Mr. Luerssen and Mr. Greene's discussion serves to answer the query of Mr. Hawkridge.

In reply to Mr. A. H. d'Arcambal's discussion the set of Fracture Standards is an empirical set of standards and cannot be produced by a definite material heat treatment formula. While this formula is essential as a guide, the fractures obtained must be discarded if they do not come up to the standard of acceptability established for each particular number. Slight variations are made in treatment during their production and a great many discards must be made for each acceptable specimen.

Another mill could use entirely different steels and heat treatments and

would accomplish the same final results if they were working to the same set of primary standards. We feel that it would be very desirable for the steel fraternity to concentrate on this set of standards, which idea was also expressed in Mr. Spalding's discussion.

It is not difficult for two observers to check the fracture factor within ½ fracture number. We ordinarily accept ¼ of a number, plus or minus, as a check, but it is not unusual for trained observers to continually check fractures to the identical reading.

All of the tool steel purchased by the Ingersoll-Rand Co. is subjected to this P-F test. We are obtaining a correlation between service performance and the P-F characteristics which will, at a later date, be transformed into a specification for P-F characteristics for the various tools.

The curves given in our heating tests represent the temperature on the inside of a piece of ¾-inch round. It is readily conceivable that the outside of a piece of ¾-inch round will heat in the short time mentioned by Mr. d'Arcambal sufficiently above the critical temperature to obtain a definite hardened case. This differential in temperature uniformity throughout the piece would also result in a faster cooling rate due to the non-necessity for the cooling medium to extract heat from the "cold" inside.

We have also definitely noted the difference in depth of hardness obtained as a function of the length of time through the critical temperature and at the quenching temperature. This resulted in the establishment of rigid hardening procedure for the conduction of the P-F test.

In reply to the discussion by Mr. A. C. Jones there is a noticeable difference in the hardening effect obtained on steel with large inclusions, as will be noticed by referring to Figs. 10, 11 and 12. Small inclusions also have an effect upon hardenability, as will be noticed from Fig. 8.

Metallographic examination of the center of each of these slabs did not show any difference in the amount of inclusions. Apparently the inclusions causing the change in hardening effect were submicroscopic in character.

In reply to the discussion by Dr. G. E. Doan may we say that we have been unable to find any correlation between the P-F characteristic and the gas content of the steel in question. Our oxygen analyses were made by the courtesy of Mr. E. C. Bain, and the nitrogen analyses were made by the courtesy of Halcomb Steel Company, and are as follows:

Sample No.	Type	Nitrogen Analysis Per Cent	Oxygen Analysis Per Cent
CA	1	0.006	0.007
AH	1	0.008	0.011
CK	1	0.008	0.012
6	3	0.007	0.013
1	3	0.10	0.014
CF	3	0.008	0.011

EFFECT OF McQUAID-EHN GRAIN-SIZE ON HARDNESS AND TOUGHNESS OF AUTOMOTIVE STEELS

By H. W. McQuaid

Abstract

This paper discusses the importance of grain-size control in automotive steels. It touches on the importance of grain-size characteristics, not only as shown by the McQuaid-Ehn test, but under other conditions as well. It discusses briefly the uncertainty of such terms as hardness, toughness, etc., as well as the underlying factors which operate to cause grain-size variation.

Examples are given of results of grain-size variation on case strength, impact value, distortion, etc. The value of the fine grain steels in making possible direct quenching of carburizing steels is noted. The application of the grain-size specification to various heat treated parts is discussed and some recommendations are made.

THE title of this paper might lead one to think that automotive steels are of a different type than those used in other industries. If by automotive steel is meant those steels commonly used in the manufacture of automotive equipment, we must cover all grades of plain carbon and all grades of alloy with the exception of the special tool steels. Automotive steels include screw machine stock, case hardening grades, bolt stock, spring steels, valve steel, oil hardening gear steels, deep drawing sheet steel, stainless, etc. The list is long and the requirements many. Because of the complexity of the requirements it will be impossible in a paper such as this to do more than touch briefly on the requirements for different parts of automotive equipment, the types used and the part played by grain-size in these types. It must be made clear at this point that the McQuaid-Ehn test is not necessarily restricted to steels for case hardening, but probably finds its most valuable application in steels of the medium carbon group both in alloy and plain carbon.

The value of grain-size control as indicated by the McQuaid-

A paper presented as part of the Grain-Size Symposium of the Sixteenth Annual Convention of the Society held in New York City the week of October 1, 1934. The author, H. W. McQuaid, a member of the Society, is research metallurgist, Republic Steel Corp., Massillon, Ohio. Manuscript received July 5, 1934.

Ehn test lies in its use to help reduce the variation in characteristics from one heat of steel to another of the same chemical specification. For many years there has existed unexplained variation in characteristics between heats of steel of the same analysis particularly in the plain carbon grades and many carbon steel applications have been replaced by alloy steel applications because of this. In most heat treating plants there have been times when lots of forgings have been found which would not respond to the quench, or which cracked in quenching. Case hardened plain carbon steel has been a particular sufferer from this variation, and in the gear plants the difficulty has been such that it has been common practice to determine the characteristics of a given heat by sending through trial on pilot lots.

The original development of the carburizing test to determine response to case hardening indicated, in the carburized samples, two distinct characteristics in a heat of steel. First, on the plain carbon steels particularly, the difference in structure of the hypereutectoid zone caused by the variation in stability of the pearlite grain in contact with the cementite network and second the variation in the size of the pearlite grain itself. See National Metals Handbook, 1933, page 644.

The variation in structure or "normality" as indicated by the McQuaid-Ehn test has been investigated by many and seems to be related to the oxide in the steel as made, and according to Grossmann also to the oxygen absorbed in the carburizing process. What makes one heat of steel absorb more oxygen during the carburizing process than another is still far from being determined but its discussion, while very interesting, has no place in this paper.

As investigations were made, however, following the original investigation, it was found that the size of the grain was of great importance, in both alloy and carbon grades, but the "normality" of more or less indeterminate importance, particularly in the alloy grades. The first tendency with the development of the so-called McQuaid-Ehn test was to consider all fine-grained abnormal types as unsatisfactory and all coarse-grained normal types as good. It was soon found, however, that this was far from being true and that the coarse grain very normal types, were not satisfactory for gears, axle shafts, bearings, etc. This lead to a study of the relative effects of variation in grain-size and normality with an increasing realization of the importance of grain-size.

The usual explanation for the variation in inherent grain-size as shown by the McQuaid-Ehn tests is that the austenitic grain-size is a function of the number and size of the nuclei present in the liquid steel at the time it freezes. Thus it is said that when the liquid steel containing a certain percentage of FeO in solution has added to it a certain amount of a strong deoxidizer like aluminum or vanadium which goes into solution in the liquid steel the deoxidizer reacts with the FeO forming nonmetallic particles which are present as finely divided solid inclusions in the liquid steel. These inclusions supposedly in colloidal dispersion act as centers of crystallization to fix the grain-size. While this is a more or less plausible explanation it is difficult to make it always fit in with actual practice. Thus we find that electric furnace heats with silicon of 0.20 per cent or more can be very easily made fine-grained while other heats with low silicon and high FeO are relatively difficult to make fine-grained. Electric furnace slags are usually very low in FeO with correspondingly low FeO in the steel. The addition of silicon and manganese before tapping should result in well deoxidized electric furnace steel yet unless the heat has been boiled down and the scrap selected, the well killed electric heat will be fine-grained with a relativly small addition of aluminum. The fact that variation in the deoxidation practice and aluminum addition affects the grain growth range as well as the grain-size, indicates that the cause of grain-size is not nuclei solid at molten steel temperature but some compound of aluminum, vanadium, etc., which is present in the austenite grain boundaries and which changes its solubility at some definite temperature in the austenite. The effect of this solubility on carbon solubility would produce normality variation, and inhibit grain growth below a given temperature. This would account for the difference in grain-size and grain growth between vanadium and aluminum killed steel and also the varying effect of different percentages of deoxidizer. Whether the compounds which produce the interference phenomenon are nitrides, carbides, or oxides, is a matter of speculation, but the indications are that the nitrides are a very important factor. The fact that grain-size control on carbon steel has been successfully accomplished commercially in plain carbon heats, by regulating carbon drop, FeO slag content, tapping temperature, aluminum additions, etc., does not necessarily mean that the FeO in the steel plus the aluminum was the final contributing combination to affect grain-size control. In heats relatively high in FeO the increased aluminum may be required

to provide an excess over and above that required for simple reaction with FeO.

Before going further we should spend some time clearing up the possible vagueness surrounding the use of the terms "McQuaid-Ehn Grain Size," "Hardness" and "Toughness."

Hardness is a rather indefinite term and while it implies a certain comparative resistance to indentation or movement of material, in a discussion involving hardness it is always best to qualify the word hardness by such terms as Rockwell, Brinell, Scleroscope, file, etc. The structure being tested has a direct bearing on the relation between file hardness, Rockwell hardness, etc. The definition of hardness as a measure of the resistance to wear does not apply to austenitic structure or structures containing a high percentage of austenite. For very hard surfaces which must resist wear and abrasion, the file, limited as it is, is still the best indicator of hardness. It is the only certain method of insuring satisfactory hardness on case hardened alloy parts quenched directly from the carburizing box. In this type of work almost any combination of file hardness and Rockwell or Scleroscope can be obtained by varying the carbon content of the case and the quenching method. It is possible to obtain by special treatment on a 2-inch round section, readings of 63 Rockwell C and a Scleroscope hardness of 100 plus on one side, with readings of 37 Rockwell C and 45 Scleroscope on the other, and at the same time have the side with the high readings soft to the file with the side with low readings file hard. In such parts as gears, the file is the best test for hardness where resistance to surface fatigue and scoring, is the important factor.

As far as toughness is concerned this is also a rather vague term and is associated in the minds of most steel users with high limits of deformation before failure. The classic example of automobile front axles and steering arms which could be tied in knots before rupture, is a good example of what the public has been taught to think of when tough steel is mentioned. The fact that the soft axle and steering arm would not stand up in service with a much harder axle, or arm was soon made evident, as the load on these parts increased. It is necessary that an important part have sufficient strength and rigidity to perform in service with minimum deflection, and toughness is not important within this range. However, when overload and deflection occurs beyond the elastic limit, it is important in most parts that there be present the ability to take a perma-

nent set without concentrating stresses on some defect in surface or design. This ability to absorb stresses beyond the elastic limit without fracture is the true useful measure of toughness. It is measured to some degree by the impact test, and must always be considered in connection with the hardness and elastic limit of the part.

The McQuaid-Ehn grain describes a certain structure obtained after carburizing at 1700 degrees Fahr. for eight hours and cooling slowly enough to develop a pearlitic structure in the case, with the grains of pearlite surrounded by excess cementite in the hypereutectoid zone to indicate their size and the character of the cementite-pearlite structure. It indicates the structure of the case and the size of the pearlite developed in the case by this temperature and slow cooling. The McQuaid-Ehn test indicates the structure only at the temperature given and does not necessarily indicate the structure or grain size at any other temperature. This is a most important point since we may have and may desire an entirely different grain size at other temperatures.

We know for good machinability in some types of steels that a coarse pearlitic structure is the most satisfactory, whereas in the same steel after heat treating, a very fine structure may be considered the most desirable. This means that in the heat treating range, a fine-grained steel would be the best and in the normalizing range, a coarse-grained steel might be the best. Thus for some applications a change in grain-size on heating is desirable, between the heat treating temperatures and the normalizing temperature.

This is fortunately what usually happens unless the steel is deliberately made to be fine-grained at 1700 degrees Fahr. when it becomes necessary to raise the normalizing temperature above the grain growth point if the best machining structure is to be obtained.

According to most text books covering the heat treatment of steel, grain growth starts at the Ac_3 point and continues at an increasing rate as we increase the temperature above the Ac_3 range. This is a condtion that only exists in the true coarse-grained steels and if the steel melting practice is modified to control grain-size (growth) we find that in some steels there is a very slight increase in grain-size until we pass a certain definite temperature which may be 500 degrees Fahr. above the Ac_3. The location of this grain growth range is a function of the deoxidation practice and is affected by the kind and amount of deoxidizers used and the condition of the liquid steel at the time they are added. The grain growth

range of the case may be different from that of the core, due to a different coarsening temperature.

In the commercial plain carbon steels, carburized at different temperature, it was found that in the fine-grained abnormal specimens (at 1700 degrees Fahr.) the core would, in every case, coarsen at temperatures below 1800 degrees Fahr. whereas the hypereutectoid zone in some cases did not coarsen below 1900 degrees Fahr.[1]

The addition of alloys and variation in deoxidizing practice results in many variations so that under certain conditions we may have quite decided lack of uniformity in relation between case and core.

Since for machinability we are interested in the structure of the material after a normalizing operation or in the as-rolled condition, we are interested in the grain-size and structure as it is found in either condition. Since most normalizing is done at temperatures close to 1700 degrees Fahr. the McQuaid-Ehn grain-size is of value for these steels.

Why some steels show a very rapid increase in grain-size at temperatures close to the Ac_3 while others show very little grain growth until temperatures as much as 500 degrees Fahr. above the Ac_3 are reached is not at the present understood. The variations in properties which can be obtained by controlling the structure and grain growth tendency is fairly well understood and of great practical importance. In order to understand why the McQuaid-Ehn test is of commercially great importance we must consider first of all what information can be obtained from it. In the first place we determine whether or not the grain shown by the test at 1700 degrees Fahr. is relatively coarse or fine. If it is coarse we know that the steel will respond better to quenching, that is, it will harden more deeply for a given quench and for the same reason warp and crack more easily if the sections are not uniform. It will for the same Brinell hardness, show lower impact values than will the fine-grained steel. It will also for the same Brinell hardness machine better with better surface than will the fine grained steel in the usual rolled or forged condition. Where minimum distortion is desirable the fine-grained shallow hardening type is desirable and since the surface hardness is little (in most carbon and alloy grades) affected by the grain size, there is a growing tendency to specify the fine-grained shallow hard-

[1]E. W. Ehn, British Iron and Steel Institute Journal No. 1, 1922, p. 170. E. C. Bain, TRANSACTIONS, American Society for Steel Treating, Dec. 1933.

ening type and increase the normalizing temperature to coarsen the grain for good machining. If the grain-size control has been carried to a point where the grain does not coarsen below 1800 degrees Fahr. it becomes difficult to put it in good condition for machining because of the scarcity of satisfactory furnaces for normalizing at temperatures above 1800 degrees Fahr.

However, many plants are now equipped to normalize the fine-grained steels at temperatures of 1800 degrees Fahr. or higher and the results have been quite satisfactory. In the alloy grades particularly the melting practice has a direct relation to the grain growth range and some alloys are particularly easy to coarsen at temperatures above 1700 degrees Fahr. The standard chromium-vanadium steels of the S.A.E. 6100 series coarsen up very easily at temperatures above 1700 degrees Fahr. and yet are usually fine-grained below 1650 degrees Fahr. This is an ideal condition and makes this steel very well suited for applications where maximum machining production is required.

Soon after the carburizing test was developed it was found that for some applications that the coarse-grained normal type was too deep hardening and warped too much so that the fine-grained type was indicated where distortion and toughness were important. The effect of grain-size as shown by the carburizing test on the properties of the heat treated part soon became known and a grain-size chart was developed to permit of classification according to grain sizes. The first chart developed was that of the United Alloy Steel Co. in 1925. It showed ten grain-sizes and was followed in a short time by the Central Steel Co. with one of nine grain-sizes. The Central Steel Co. chart included not only the hypereutectoid zone at 1700 degrees Fahr. but the gradation zone and core as well. Today the American Society for Testing Materials grain-size chart is commonly used and can be considered as standard.

Development of the grain-size chart was soon followed by specifications for grain size, particularly on case hardened alloy gear steels. The tendency to specify according to grain size soon required the steel mills to develop a technique which would enable them to meet the specified grain-size requirements. The alloy mills were soon forced by the increasing pressure to study melting practice from scrap to ingot with the greatest of care, and satisfactory control was finally obtained to the point where fine or coarse-grained steels could be made commercially. Certain alloy combinations present difficulty

and do not lend themselves to grain-size control. The plain carbon open-hearth heats are somewhat difficult to control and require as a rule, relatively more care and skill than alloy heats of the same carbon and manganese content.

The development of grain-size control on medium carbon automotive steels is particularly interesting in that in these steels the hardening characteristics are of especial importance. With some of the largest users, a steel of approximately 0.40 per cent carbon and 0.80 per cent manganese is used for front axles, cranks, cams, rocker arms, starter rings, connecting rods, etc. Here, while the analysis is identical, the hardness requirements are of necessity different and the steel maker must supply one type for connecting rods, another type for cranks and still another for front axles or starter rings. This variation in requirements for hardenability is met with commercially by grain-size control as indicated by the McQuaid-Ehn test and since the ability of the steel supplier to hold this large tonnage business depended on this ability to meet the requirements, a sudden and intense interest in grain control was developed when this became known.

Thus, we find today, all the important producers of special carbon as well as alloy automotive steels for heat treatment in a position to supply these steels to different types. The development of the ability to do this has marked a very important and revolutionary change in the view point of the large scale steel producers.

The ability to produce steels which retained their fine-grained characteristics after prolonged heating at carburizing temperatures, gave also a new importance to quenching directly from the carburizing box and temperature. Since the usual steel was formerly coarse-grained, quenching directly from the carburizing temperature was in ill repute because of the poor fracture and high warpage obtained. With the advent of fine-grained steels, however, it became possible to quench directly from the box and still have a satisfactory fracture combined with good toughness and low warpage. This has resulted in a considerable increase in the number of carburized parts which are directly quenched from the carburizing box.

In considering the effect of grain-size variation from heat to heat as shown by the McQuaid-Ehn test on finished properties we should be careful to deal with averages and not be too quickly misled by special cases. As shown by Bain and others, the fine-grained steels are usually shallower hardening and have higher impact values

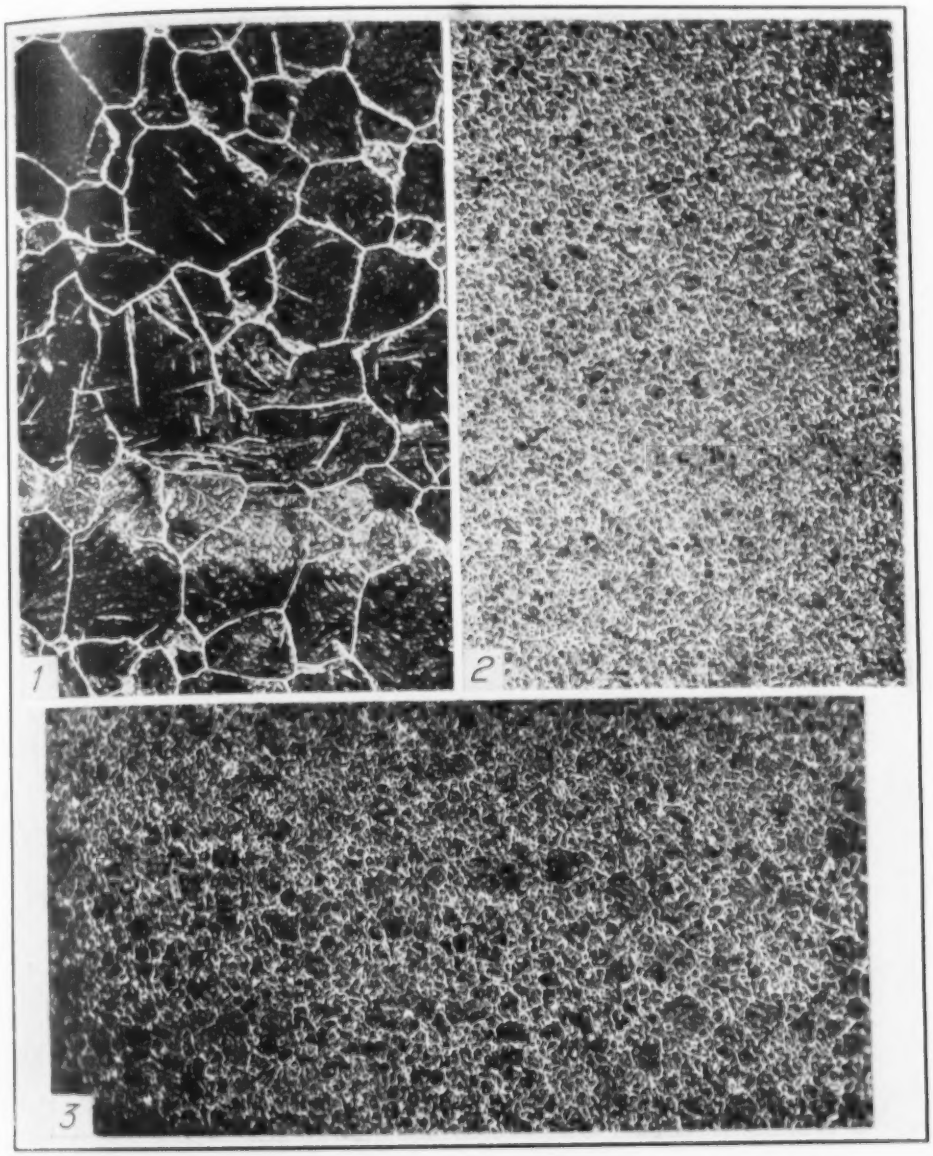


Fig. 1—Coarse Grain Normal Type, With Abnormal Streak. × 100.
Fig. 2—Fine Grain Abnormal Type. × 100.
Fig. 3—Fine Grain Normal Type. × 100.
Photomicrographs courtesy E. V. Blanchard, Bower Roller Bearing Co.

than the coarse-grained steels. Exceptions to this have been noted many times, particularly in the alloy grade. The so-called fine-grained normal types are apt to react differently from the fine-grained abnormal type and in the alloy carburizing steels the fine grain normal heats tend to break in cold straightening and often show most of the undesirable properties of the coarse normal heats. The fine grain

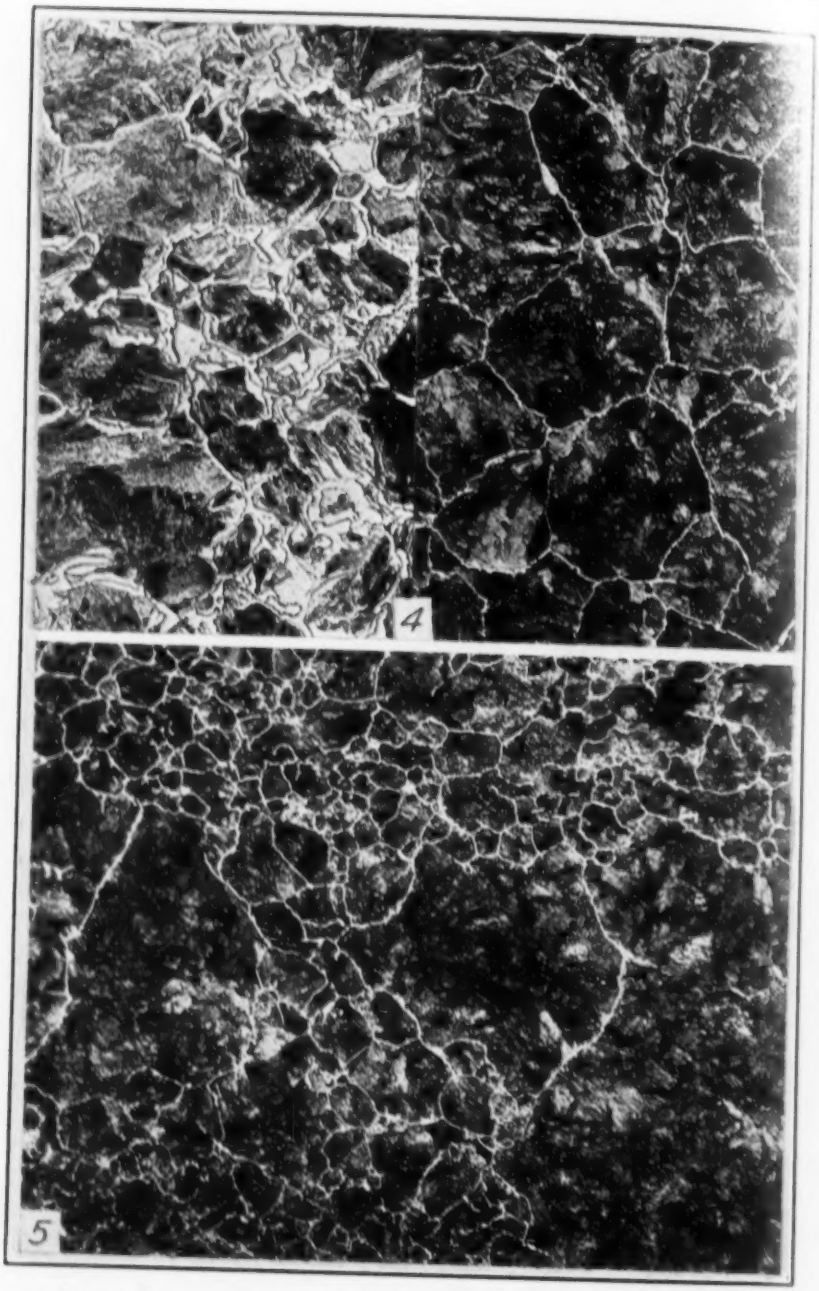


Fig. 4—Photomicrograph at Left is that of a Coarse Grain Abnormal Steel. × 100. Photomicrograph at Right is that of a Coarse Grain Normal Steel. × 100.

Fig. 5—Photomicrograph of Mixed Grained Normal Type Steel. × 100.

normal heats do not harden as deeply as the coarse grain normal heats but they do show a tendency (in the alloy steels particularly) to be less ductile and more sensitive to suddenly applied loads than

the fine-grained abnormal types. For this reason, where the maximum combination of toughness and hardness is required, the fine-grained abnormal type is to be preferred.

Cases have been found where (in the high carbon plain carbon steels) the coarse-grained abnormal types have shown shallow hardening properties. In checking tool steels it has been noted that some of the coarse-grained abnormal types show good resistance to grain growth and would be rated as having a good timbre test.

Usually the grain-size and normality are related. The coarse grain steels show a well defined net work of cementite surrounding a finely lamellar pearlite in the hypereutectoid zone while the fine grain steels show a tendency toward massive cementite with some divorce of the pearlite. In looking back over many investigations of difficulty with heat treatment, it has been found as a rule, that where hardenability or brittleness has been involved that the fine-grained steels were shallower hardening and the coarse-grained steels were deeper hardening steels. For this reason the characteristics of the fine-grained abnormal type are the characteristics usually associated with fine-grained steel and the characteristics of the coarse-grained normal type are those usually associated with coarse-grained steel. Apparently the grain-size effects outweigh the normality effects whatever they may be.

The McQuaid-Ehn grain-size is specified by most automotive metallurgists either directly or indirectly, because of its bearing on machining, on distortion or on actual physical properties. Thus one user for years of S.A.E. 2315 ring gear steel would specify a coarse-grained steel (1-5 on the American Society for Testing Material's chart) because with his plant procedure this gave him the best results in machining, but at the expense of the distortion in the hardening.

Another user would specify a fine-grained steel (6-8 on the American Society for Testing Material's chart) for the same part because of reduced distortion as compared to the coarse-grained steel. By normalizing at a high temperature the rate of machining was not reduced commercially. Some users rather than change to a higher alloy or carbon analysis will specify a coarser grain to improve the hardenability.

Most gear applications require a fine grain, shallow hardening type because in gears the control of distortion is usually of the utmost importance to avoid heavy localized tooth pressures and noise by irregular and small tooth contacts. This is quite universally the case

in automotive applications where gears are usually designed with a low factor of safety and their success depends on good tooth contacts.

Some users, especially on plain carbon steel, require a steel which shows an intermediate range of grain-size on the McQuaid-Ehn test. This is to permit satisfactory hardenability on heavy sections without cracking on the lighter sections. The tendency to crack is much greater on the coarse-grained, deep hardening steels so that in the water-quenched carbon steels the coarse grain must be avoided.

In oil-hardened, plain carbon gears, such as flywheel starter rings, it is necessary to obtain maximum hardenability especially in the lower carbon grades. Here the coarser-grained steels are required and perform quite satisfactorily while the finer-grained shallow hardening type are generally entirely unsuitable.

In connecting rods, on the other hand, the coarse-grained, deep-hardening type will not work out satisfactorily because of their tendency to be more sensitive to shearing, forging and flash line cracks. The sections are thin so that in order to avoid high heat treat losses the shallow hardening grades are required.

Axle shafts are generally of the fine-grained type particularly in heavy duty truck applications. The combination of relatively high alloy and fine grain permits the use of high Brinell hardness with its consequent good fatigue life, together with the decreased tendency to start localized cracking of the fine-grained types. These surface cracks act as notches for fatigue failures and are to be avoided whenever possible. The use of fine-grained steels is for this purpose, particularly important. The greater tendency of the coarse-grained, deep-hardening types to localize severe stresses, results in much shorter life on heavy duty axle shaft applications due to the occasional heavy overstress to which this member is subjected. On passenger car service the overstresses are much fewer so that lower alloy, coarser-grained steels are satisfactory.

In the plain carbon, case hardened parts, such as camshafts, the coarse-grained steel is preferred because of its easier quenching to avoid soft spots. The difficulty encountered with the very coarse camshafts is a tendency to quench so hard that the corners of the cams crack off. For this reason a finer-grained type which is still satisfactory for good surface is to be preferred. For plain carbon piston pins, the same requirements are present as in camshafts. In alloy pins the fine-grained grades are much to be preferred. The

coarse-grained types are more sensitive to grinding cracks and are undesirable from this standpoint.

For such parts as knuckle pins, steering arm balls, etc., the fine-grained shallow hardening type is to be preferred because of the importance of these parts in the safe operation of the vehicle. The wider heat treating range, the superior impact properties and the

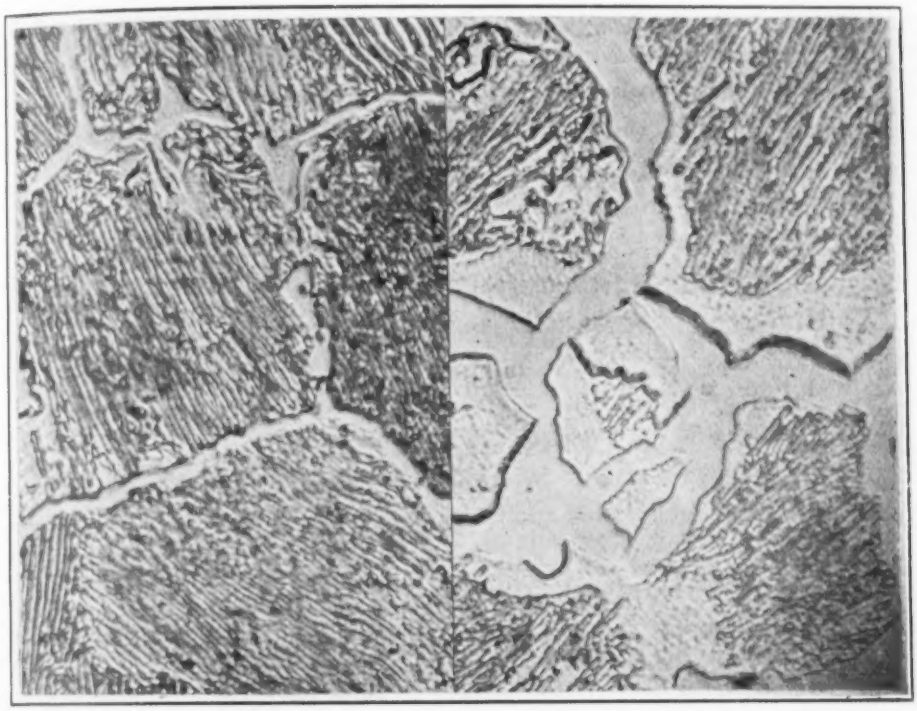


Fig. 6—Photomicrograph at Left is that of a Fine Grain Normal Type Steel. × 1000.
Fig. 6—Photomicrograph at Right is that of a Fine Grain Abnormal Type Steel. × 1000.

generally tougher characteristics of the shallow hardening type indicate the use of the fine-grained type. The same argument can be used in favor of the fine-grained steels for steering arms, knuckles, steering gear parts, etc. These parts should always be made and treated to give the maximum of notch fracture resistance, and toughness for the hardness required.

For cold-headed bolts and in fact all heat treated bolts whether alloy or carbon, water- or oil-hardened, a fine-grained, shallow hardening type is to be preferred. However, for cold heading, a steel which will be very coarsely pearlitic in the as-rolled, or nomalized condition is to be preferred. This is because of the better cold

heading properties of the steel which shows a very coarse pearlitic structure. This means that if a fine-grained steel is used in the normalized condition, the normalizing operation should be such as to coarsen the pearlite as much as possible and if used in the as-rolled condition that the finishing rolling temperature should be as high as possible, and above the coarsening range.

For springs, the finer-grained steels with their decreased tendency to localize stresses at surface defects are to be preferred. Due to the high quenching stresses still present in the finished spring, the tendency to localize stresses is very pronounced. In flat springs this is offset by the fact that most surface defects due to rolling, etc., are in such a position as to not play an important part. With the advent of the coil spring the effect of surface defects was very much increased, because of the torsional stresses involved. Hence the fine-grained shallow-hardening types are to be preferred. In all such applications, where the quench is severe the control of the depth of hardening is important and hence grain-size as an indicator of hardenability is important.

It is possibly simpler and more accurate in such applications to actually make grain growth and depth hardness checks using a method of standardization such as that developed by B. F. Shepherd of the Ingersoll-Rand Co. This method is well suited to the high carbon steels and should be of advantage on such steels as the silico-manganese types.

In steel for automotive bearings of the roller type the finer-grained steels are almost universally used. These are usually of the nickel-molybdenum carburizing grade and are carefully checked for grain size. The fine-grained steel in this application is of value primarily because of reduced distortion, decreased internal stresses from the quench and less tendency to check in grinding.

In heat treated steels for miscellaneous parts such as hardened washers, small pins, etc., of plain carbon steel, usually cyanided and sometimes case hardened, grain-size and normality checks are not often used because of the difficulty in controlling the steel for such parts at the source. Where cyaniding is to be used, the fine-grained shallow-hardening type is to be preferred especially for washers, because of the difficulty with distortion after a water quench. For parts not heat treated, such as frames, body and fender stock, the McQuaid-Ehn grain-size is important but not specified because of the greater importance of other factors. In the making of the steel

Table I
Effect on Strength in Tension of Carburized Steel Due to Grain-Size Variation
Carburized 1700 Degrees Fahr. Pot Quench. Drawn 325 Degrees Fahr. One Hour.
Average Values

Steel	Case Strength First Crack		Case Depth		Deflection	
	Fine	Coarse	Fine	Coarse	Fine	Coarse
	Pounds per Square Inch		Inches		Inches	
1020	155,500	138,500	0.030	0.034	0.086	0.060
1020	158,300	113,000	0.045	0.048	0.077	0.048
1020	143,700	95,700	0.065	0.064	0.066	0.036
1020	146,300	113,000	0.072	0.074	0.072	0.055
2315	258,000	212,000	0.030	0.035	0.159	0.121
2315	205,000	180,000	0.054	0.045	0.112	0.099
2315	219,000	165,000	0.065	0.062	0.122	0.090
2315	210,700	164,000	0.072	0.072	0.124	0.085
4615	192,700	179,000	0.027	0.028	0.122	0.092
4615	191,000	148,000	0.045	0.045	0.105	0.074
4615	191,700	149,700	0.060	0.066	0.105	0.075
4615	203,200	145,000	0.078	0.074	0.110	0.075
2512	253,300	253,000	0.031	0.032	0.162	0.161
2512	225,600	203,000	0.045	0.048	0.150	0.126
2512	267,600	187,800	0.060	0.060	0.115	0.104
2512	176,300	193,000	0.075	0.078	0.096	0.114
3115	187,900	174,400	0.030	0.030	0.099	0.099
3115	178,700	158,400	0.048	0.046	0.092	0.083
3115	165,000	155,500	0.066	0.064	0.080	0.078
3115	165,800	155,000	0.080	0.076	0.083	0.077
6115	191,200		0.028		0.114	
6115	172,200		0.050		0.090	
6115	170,900		0.066		0.082	
6115	158,400		0.080		0.080	

for deep drawing sheets, the practice required where surface is all important, interferes greatly with McQuaid-Ehn grain-size control.

Some examples are herewith given of grain-size effects on parts of an automotive vehicle together with a table showing the usual steel and grain-size for such parts.

Following are some extracts from reports of actual investigations of actual plant difficulties, where grain size was involved.

Investigation of High Rejections of Gears Because of Noise

(1) "The follow up data on this heat is indicative that the fine-grained type is less susceptible to movement than the coarse-grained type, therefore, changing the grain-size specification on the material for the gear from the coarse to the fine side was justified."

Investigation of Losses Due to Distortion of Pinion Stems in Quenching

(2) "From the above it is apparent that, since there is a varia-

Table II
Effect on Strength in Tension of Carburized Steel Due to Grain-Size Variation
Carburized 1700 Degrees Fahr.—Single Treatment—Drawn 325 Degrees Fahr. One Hour.
Average Values

Steel	Case Strength First Crack		Case Depth		Deflection		Treatment
	Fine	Coarse	Fine	Coarse	Fine	Coarse	
	Pounds per Square Inch		Inches		Inches		Degrees Fahr.
1020	137,800	158,000	0.030	0.030	0.090	0.094	1425 Water
1020	137,000	137,700	0.055	0.058	0.074	0.066	
1020	163,000	166,000	0.070	0.075	0.090	0.084	
1020	157,000	146,000	0.078	0.082	0.079	0.069	
2315	226,700	182,300	0.040	0.038	0.134	0.102	1475 Oil
2315	228,300	167,000	0.055	0.058	0.130	0.085	
2315	252,000	173,000	0.064	0.064	0.163	0.089	
2315	222,000	155,500	0.078	0.078	0.123	0.076	
4615	168,000	148,000	0.038	0.036	0.100	0.075	1500 Oil
4615	182,000	129,000	0.048	0.053	0.107	0.060	
4615	189,000	136,000	0.064	0.060	0.100	0.065	
4615	199,000	137,300	0.075	0.080	0.111	0.063	
2512	275,100	242,800	0.032	0.028	0.184	0.146	1450 Oil
2512	237,100	257,400	0.050	0.055	0.148	0.159	
2512	258,300	250,100	0.070	0.065	0.173	0.148	
2512	239,600	244,100	0.080	0.075	0.139	0.146	
3115	164,200	140,100	0.038	0.036	0.086	0.074	1475 Oil
3115	148,700	152,800	0.058	0.052	0.071	0.080	
3115	167,500	162,000	0.065	0.065	0.083	0.085	
3115	185,000	165,000	0.078	0.080	0.092	0.083	
6115	200,400		0.039		0.115		1625 Oil
6115	210,700		0.056		0.129		
6115	192,800		0.065		0.105		
6115	208,500		0.080		0.113		

tion of five (5) numbers in the grain-size ratings, of the pinions, there is, at least in this particular instance, a definite relation between grain-size and pinion stem straightness. It is also apparent that a coarse-grained type should not be used when direct quenching is employed."

The above is typical of many reports which could be quoted to show the importance of grain-size and is indicative of the practical application of grain-size control.

Tables I, II and III show the effect on the strength in tension of a carburized case due to grain-size variation. The values were determined by bend tests of ¾-inch square bars, 10 inches long, carburized on top and bottom only. The figures for case strength represent the calculated values obtained from the applied bending load. These figures show not only the advantage of the fine-grained case but they also show the weakening of the case with increase in depth. This decrease in strength with increasing depth is most marked in

Table III
Effect on Strength in Tension of Carburized Steel Due to Grain-Size Variation

Carburized 1700 Degrees Fahr.—Double Treatment—Drawn 325 Degrees Fahr. One Hour.
Average Values

Steel	Case Strength First Crack		Case Depth		Deflection		Treatment
	Fine	Coarse	Fine	Coarse	Fine	Coarse	
	Pounds per Square Inch		Inches		Inches		Degrees Fahr.
1020	133,000	143,700	0.030	0.037	0.090	0.095	1625 Oil
1020	147,000	131,700	0.055	0.062	0.092	0.068	1425 Water
1020	155,700	165,500	0.066	0.072	0.089	0.090	
1020	152,300	160,700	0.074	0.078	0.092	0.084	
2315	260,000	227,500	0.038	0.035	0.163	0.153	1550 Oil
2315	257,300	235,700	0.064	0.055	0.170	0.178	1380 Oil
2315	296,000	236,000	0.068	0.065	0.233	0.163	
2315	293,500	259,000	0.078	0.085	0.204	0.180	
4615	173,400	155,300	0.035	0.040	0.114	0.090	1550 Oil
4615	180,000	177,000	0.048	0.050	0.125	0.104	1400 Oil
4615	200,000	174,000	0.064	0.065	0.130	0.198	
4615	197,400	192,700	0.078	0.080	0.116	0.106	
2512	303,000	254,500	0.036	0.027	0.223	0.158	1550 Oil
2512	296,900	314,700	0.052	0.048	0.210	0.235	1450 Oil
2512	311,100	298,000	0.070	0.065	0.216	0.202	
2512	309,200	286,300	0.080	0.078	0.213	0.173	
3115	179,200	132,400	0.038	0.032	0.114	0.086	1550 Oil
3115	192,000	164,950	0.052	0.048	0.114	0.114	1425 Oil
3115	192,700	157,000	0.065	0.060	0.109	0.093	
3115	197,200	181,200	0.075	0.075	0.104	0.108	
6115	170,900		0.034		0.128		1650 Oil
6115	170,100		0.056		0.096		1475 Oil
6115	196,000		0.060		0.119		
6115	202,900		0.078		0.121		

Table IV
Effect of Grain-Size on Izod Impact Value
Medium Carbon Steel

Steel	Per Cent		Grain Size
	Carbon	Manganese	
A	0.40	0.85	Fine Grain #7-8
B	0.40	0.80	Coarse Grain #1-2

Quenched 1500 Degrees Fahr.—Water
Drawn as Shown

Izod Impact Values and Brinell

Degrees Fahr.	A		B	
	Impact Value	Brinell	Impact Value	Brinell
Draw 600	48	242	9	267
800	62	234	10	234
1000	77	207	15	212
1200	98	190	33	184

the straight nickel steels and is probably due to the increasing percentage of austenite in the case. As the austenite layer increases in depth, the internal stress should increase due to the tension set up in the austenite by the martensite layer beneath it.

That the decrease in strength is probably due to increasing

Table V
Showing Commonly Used Automotive Steels and usual McQuaid-Ehn Grain-Size Requirements

	Steel	Grain Size Recommended
(a)	Plain carbon not heat treated parts. S.A.E. 1112 S.A.E. 1010 S.A.E. 1020 etc. S.A.E. X-1315	Coarse for maximum machinability and best cold heading properties. Fine when subjected to shock loads or occasional severe over-stress.
(b)	Plain carbon heat treated uncarburized parts. S.A.E. 1025 S.A.E. 1040 S.A.E. 1050 etc. Ford EE, EEE, etc.	Generally fine-grained, shallow-hardening type. When oil-hardened to high Brinell use coarse grain type. Coarse grain in very heavy parts such as large truck front axles. Medium grain in passenger car front axles.
(c)	Plain carbon case hardening type. S.A.E. 1010 S.A.E. 1020 S.A.E. X-1315	Fine grain for minimum distortion. Coarse for machining and hardening. Generally recommended fine as possible consistent with satisfactory surface hardness. Used in camshafts, small gears, some differential parts, brake parts, washers, etc.
(d)	Plain carbon high carbon type. S.A.E. 1060 S.A.E. 1095 etc.	Flat springs, valve springs, coil springs. Fine grain best to avoid stress concentration. Bumper bar stock medium grain size to coarse for oil hardening.
(e)	Plain carbon, high manganese with or without high sulphur. S.A.E. 1340 S.A.E. 1360 etc.	Used in many important parts. Fine grain preferred. Depends on analysis, quench, etc. Not commonly specified.
(f)	Low carbon, low alloy case hardening steels. S.A.E. 2015 S.A.E. 2115 etc. Chromium-Nickel Vanadium-Gear Steel C.N.V.	Fine grain for gears—Danger of trouble with surface soft spots if too fine and abnormal. For heavy parts, coarse grain. For direct quenching, fine grain preferred.
(g)	Low carbon, medium alloy types. S.A.E. 3115 S.A.E. 6115 S.A.E. 4615 S.A.E. 2315 etc. carburizing grade	For gears and bearings, fine grained best. Fine grain for direct quench. Coarse grain for heavy sections for maximum core properties. For maximum machinability with increased distortion coarse grain preferred unless special normalized.
(h)	Low carbon, high alloy type. Case hardening 3½% Ni. Molybdenum S.A.E. 2512 S.A.E. 3300 S.A.E. 3400 Krupp etc.	Used for truck and bus gears and very heavy duty case hardened parts. Fine grain in all applications.
(i)	Medium carbon, medium alloy. S.A.E. 3130-3140 S.A.E. 4130-4140 S.A.E. 5130-5140-5150 S.A.E. 6130-6140-6150	Used for passenger car and light truck axle shafts, arms, knuckles, some connecting rods, cranks, oil-hardened gears, etc. Should be fine-grained in all applications where so called toughness is important. In heavy sections where deep quench is required medium or coarse grain is best.
(j)	Medium carbon, high alloy steel. S.A.E. 3240 S.A.E. 3330-3340 S.A.E. 3335 S.A.E. 3440 Cr-Ni-Molybdenum etc.	Used for heavy duty gears, axle shafts, highly stressed parts. Always fine grain when used for above parts.

austenite is also indicated by the fact that the double treatment generally shows a stronger case than does the direct quench. With the double treatment the case is quenched at a relatively low temperature which would produce the minimum of austenite as compared to the direct quench. The increasing strength with increasing depth as shown by the double treated steels indicates the normal effect which should be producted if no increasing internal stresses were developed. It is interesting to note that in these tests that the fine-grained cases showed improved properties as compared to the coarse grain type, probably due to the inherent tendency for the coarse-grained type to localize stresses, which in material of such low ductility, will result in variable and generally poorer properties as compared to the fine-grained type.

These results would indicate that for maximum case strength, a double or single treated fine-grained case is to be preferred to the pot-quenched, fine- or coarse-grained. The effect of carbon content, while not shown, is an important factor and increasing carbon, above eutectoid, generally results in decreased case strength, so that for maximum case strength the combination of eutectoid carbon, fine-grained, and martensitic case is indicated.

Table IV shows the effect of the grain-size on the impact properties of two medium carbon, water hardening steels. These were taken from commercial heats of steel which were water quenched to a high hardness. Steel B which was one of the coarse-grained type shows very high losses due to quenching cracks and could not be used commercially. Steel A gave no trouble and, when substituted for steel B, eliminated the losses due to quenching cracks. It should be noted particularly that with an identical Brinell hardness of 234, the Izod values on Steel A were approximately six times that of Steel B. This grade of steel is particularly sensitive to grain-size effects and since it constitutes a very large tonnage in automotive steels which are heat treated, the control of grain-size becomes of very great commercial importance.

We must, in conclusion, warn against the tendency to use grain-size effects and steel-making variation as a scape goat for all variations in heat treatment or machinability. The fact remains that uniform conditions in heat treating and processing a lot of steel are just as important as ever, and when difficulties arise, a check up of the heat treat department and other related processes is just as much in order as it was before the advent of the McQuaid-Ehn test.

DISCUSSION

Written Discussion: By Philip Schane, Jr., Duquesne Works, Carnegie Steel Co., Pittsburgh.

In reading Mr. McQuaid's paper, it was interesting to note that he has concluded that the cause of grain growth restriction is not a dispersion of alumina or other refractory particles, as was formerly generally assumed. I have also been of the same opinion for some time, due to the many observed phenomena which cannot be satisfactorily explained by the dispersion theory, even when it is assumed that the dispersion is the product of an equilibrium reaction which takes place during cooling or solidification.

Some years ago, it was found that tungsten, to which a small amount of thoria was added, when made into lamp filaments was not subject to "offsetting." This "offsetting" was caused by a coarse-grain structure, one grain in some cases comprising the entire cross section of the filament. The reason for the better results with the thoriated filament was the finer grain structure obtained. Sometime later, it was found that these thoriated filaments, when given certain thermal treatments, had a much higher rate of electron emission. Dr. Langmuir, in published articles at that time, showed by mathematical calculation and laws of physical chemistry that the electron emission obtained was that which would occur if the surface was covered with a layer of thorium one atom thick and he concluded that such was the case. To obtain the higher electron emission, it was necessary to reduce some of the thoria by heating the filament to a high temperature (2800° K) and then diffuse a layer of thorium to the surface by heating at a lower temperature (2100° K).

When the cause of grain growth restriction was first investigated, it was concluded that the thoria particles were responsible for the grain-size in the tungsten and that alumina (or other refractory material) particles caused grain-size restriction in steels. The second of the above investigations, with reference to the surface film of thorium, was evidently not considered, although it is only necessary to change the unit of mass from the complete filament to a single grain in order to consider the filament as made up of grains of tungsten, each of which was enclosed by a film of thorium, which film was effective in preventing grain growth until a certain high temperature was attained.

There are a number of observed phenomena which are difficult or impossible to explain satisfactorily by the dispersion theory, such as the changes of grain growth temperatures and grain-sizes with varying metalloid and alloy contents of steels, the effects of preliminary working or treatment, etc. The following is a statement of a theory of grain structure which seems to provide a satisfactory explanation:

When steel is heated to temperatures above its critical range, recrystallization takes place and it is composed of fine-grained austenite. This austenite contains certain substances (elements, compounds, oxides, etc.) whose solubilities and rates of diffusion in the austenite vary with temperature. Under certain conditions, these substances diffuse or are rejected to the grain boundaries and form intergranular films, enclosing the austenite grain and thus inhibiting grain growth until such a temperature is attained as will cause ab-

sorption or rupture of this film and allow the grains of austenite to come in contact with each other and grow to the size normal for the temperature.

The above theory is also very useful in explaining many other phenomena at high temperatures, such as "red shortness" in certain steels (copper, sulphur, etc.), critical temperature ranges in which steel cannot be successfully worked without cracking or other surface defects, greater plasticities of certain steels at forging temperatures, variations in machinability after certain treatments, normality, etc. With slight modifications to include low temperature structures, explanations are possible for blue brittleness, aging, etc. Space is not available in this discussion for details.

The wide application of this theory to the explanation of experimental and observed data is very interesting. The evidence for it has been purely circumstantial. Further experimental work must be done to establish it.

M. A. GROSSMANN:[1] I am tempted to mention one point in regard to grain growth and hypereutectoid cementite which Mr. McQuaid brought up. Of course the interference with grain growth by hypereutectoid cementite can take place only in steels which are hypereutectoid before the McQuaid-Ehn test, as when a low carbon steel is carburized at 1700 degrees Fahr. for a number of hours, the hypereutectoid cementite which we see when the sample is cooled down is of course in solution at the carburizing temperature and does not offer resistance to grain growth.

Author's Reply

I appreciate very much the comments of Mr. Schane since they indicate that there are others who feel that the usual explanations of grain genesis and growth do not completely cover the existing facts. While it is difficult, of course, at the present time to do more than speculate on the reasons as to the variable grain growth ranges in different heats of steel, it is at least interesting, and no doubt this field will be subjected to some very interesting research in the near future.

In regard to Dr. Grossmann's comment as to the effect of the hypereutectoid zone on grain growth, it might be stated that the point which he brings out as to the uniform solution of carbon in the austenite at temperatures above the Ac_3 has been often stated, and if we are to believe the text books, is, without a doubt, true. It was merely intended to bring out that in spite of this commonly accepted assumption that the carbon is uniformly distributed in the austenite in the carburizing temperature range and therefore should not affect the grain-size on cooling, that the hypereutectoid zone in the plain carbon steels resisted coarsening at a range considerably above that of the core or gradation zone. Since the important difference between the hypereutectoid zone and the core is in the percentage of carbon, it is felt that the carbon content is directly or indirectly a factor in restricting grain growth in the fine-grained steel.

---

[1]Research metallurgist, Illinois Steel Co., South Chicago, Ill.

EFFECTS OF McQUAID-EHN GRAIN-SIZE ON THE STRUCTURE AND PROPERTIES OF STEEL

By Philip Schane, Jr.

Abstract

This paper discusses generally the different characteristics and properties of grain size controlled steels.

A more particular discussion is made of S.A.E. 1040 forging steels and data are presented to illustrate the widely different characteristics and properties of steels of inherently coarse and fine grain types.

The influence of mixed grain on the physical properties and the desirability of a uniform structure at the heat treating temperatures, which can only be obtained by a control of grain-size in steel manufacture, is clearly demonstrated.

THIS symposium on "Grain Size" is significant of the importance now attached to this extremely interesting phase of steel metallurgy by both consumers and makers alike. One conversant in a general way with the steel arts might be pardoned for inquiring why greater interest and special effort had not been directed to this subject before. This question might be answered in general by referring to the fact that competition has become increasingly keen among steel users in all lines and demands for better quality at lower costs have raised quality standards and increased manufacturing tempo. The speed and requirements of many operations have been increased to such extents that the raw materials must be uniform in quality within very narrow limits. Operations are no longer planned nor parts designed to permit the use of occasional heats which might come outside these restricted ranges.

In order to meet the increasing demands for greater uniformity, steel manufacturers have done a vast amount of investigational work of the many details of manufacturing practices and have made much progress, especially in control of grain growth tendencies in the steel

A paper presented as part of the Grain-Size Symposium of the Sixteenth Annual Convention of the Society held in New York City the week of October 1, 1934. The author, Philip Schane, Jr., a member of the Society, is Metallurgical Engineer of the Duquesne Works of the Carnegie Steel Co., Pittsburgh. Manuscript received September 6, 1934.

making process and in adapting these controlled steels to the consumer's uses and requirements.

In the steel making process, the control of grain size tendencies and attendant properties have been thoroughly investigated, and the effects of the various practices have been determined. As is now generally known, grain growth tendencies are inherent in the steel ingot when cast and are controlled in the steel making process by proper practices, deoxidation and additions. At the present time, the means to obtain the desired characteristics and uniformity of quality are known and the more progressive steel companies have large corps of metallurgically trained men to control these practices and also to develop and investigate new steels.

The demands of the customers have also made necessary extensive investigations by steel makers in order to determine the properties of these controlled steels and adapt them to their customer's use. Progress has been made to such an extent that, with a knowledge of the customer's process and requirements, a steel can be furnished which will be the optimum for the purpose, unless the metallurgist is hampered by obsolete or unreasonably close chemical or physical specifications, which were arrived at by empirical means.

It was only natural, when grain-size of steels was first considered, that the McQuaid-Ehn carburizing test should be used to classify the steels, as it was at the time the only test recognized for the purpose of determining grain-size and normality. However, as further progress was made and more complete knowledge obtained regarding the mechanics of grain growth and its control, it became evident that, while this test was a useful tool, it did not furnish sufficient information to fully define the grain growth characteristics and properties of the steel. In investigational work, other tests, usually for grain-size and hardenability of the uncarburized sample heat treated at various temperatures, are being used.

A large amount of experimental and service data has been obtained on these controlled steels in various grades, compositions and charactertistics, but, in order to illustrate the differences which exist between steels of practically the same chemical composition but of different grain growth tendencies, definite data are being submitted here on S.A.E. 1040 steels of the coarse- and fine-grained types classified by the McQuaid-Ehn test. This particular composition has been selected for an example on account of its general use in forging and heat treating processes.

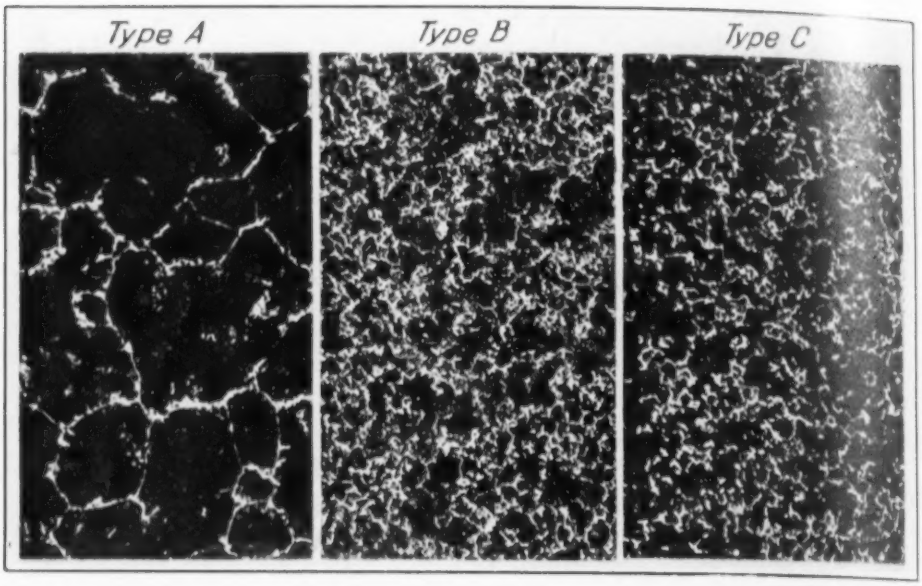


Fig. 1—Representative Grain Structures of Carburized Tests of Three Types of Steel, Standard McQuaid-Ehn Procedure Being Used. Types Include Coarse- and Fine-Grained of Approximately the Same Composition and a Fine-Grained Type of Lower Carbon and Higher Manganese.

In Fig. 1 are shown representative grain structures of carburized tests of three types, standard McQuaid-Ehn procedure being used. The types include coarse- and fine-grained of approximately the same composition and a fine grained type of lower carbon and higher manganese.

Fig. 2 shows the microstructure of 1⅛-inch rounds of coarse- and fine-grained steels as rolled, the finishing temperature being approximately the same in both cases (about 1800 degrees Fahr.).

In Fig. 3 are shown composite micrographs showing the actual grain-sizes of the coarse- and fine-grained types after normalizing at various temperatures from 1400 degrees Fahr. to 2000 degrees Fahr. (760-1095 degrees Cent.). Normalizing was for one hour at the temperatures shown, the increment of temperature being 50 degrees Fahr.

In Fig. 4 are shown the fractures of 1-inch rounds of coarse- and fine-grained types, of the same heats as Figs. 2 and 3, as quenched from 1400 degrees Fahr. to 2000 degrees Fahr. (760-1095 degrees Cent.). Fractures for each 50 degrees Fahr. temperature difference are shown.

To further illustrate the differences in hardenability, hardenability curves are shown in Fig. 5 of coarse- and fine-grained steels

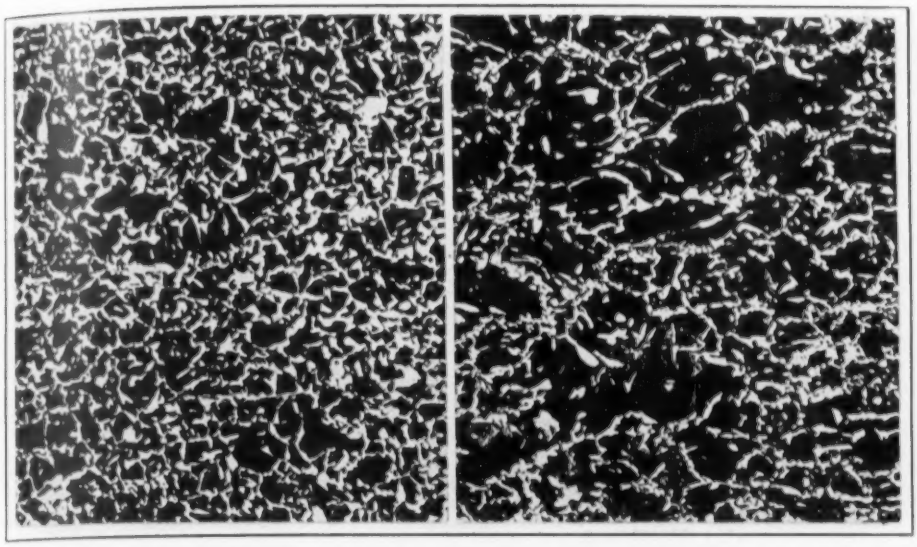


Fig. 2—Microstructure of 1⅛-Inch Rounds of Coarse- and Fine-Grained Steels as Rolled, the Finishing Temperature Being Approximately the Same in Both Cases (About 1800 Degrees Fahr.).

of the same composition (same heats shown in Figs. 2, 3 and 4) as quenched from various temperatures in the ordinary heat treating range. These show the slight effect on the fine-grained type of wide variations in the ordinary range of quenching temperatures, as compared with the much greater effects of the same differences when coarse grain steel is used.

Fig. 6 and Table I show in graphical and tabular forms the comparative physical properties of the coarse- and fine-grained types of S.A.E. 1040 steel after normalizing for one hour at temperatures from 1400 to 2000 degrees Fahr. (760-1095 degrees Cent.) using 100 degrees Fahr. intervals.

Fig. 7 and Table II show the microstructure (100 X) and physical properties of the coarse- and fine-grained types of S. A. E. 1040 steel after annealing in an electric furnace. The temperature of the steel was 1300 degrees Fahr. (705 degrees Cent.) for fourteen hours, followed by four hours furnace cooling at 50 degrees Fahr. per hour. The cooling was finished outside the furnace in a controlled atmosphere so that at the end of thirty hours the bars were cool enough to handle. It is interesting to observe the finer structure of the fine grain steel, its greater softness and higher impact values.

Fig. 8 and Table III show in graphical and tabular forms the comparative physical properties of the coarse- and fine-grained

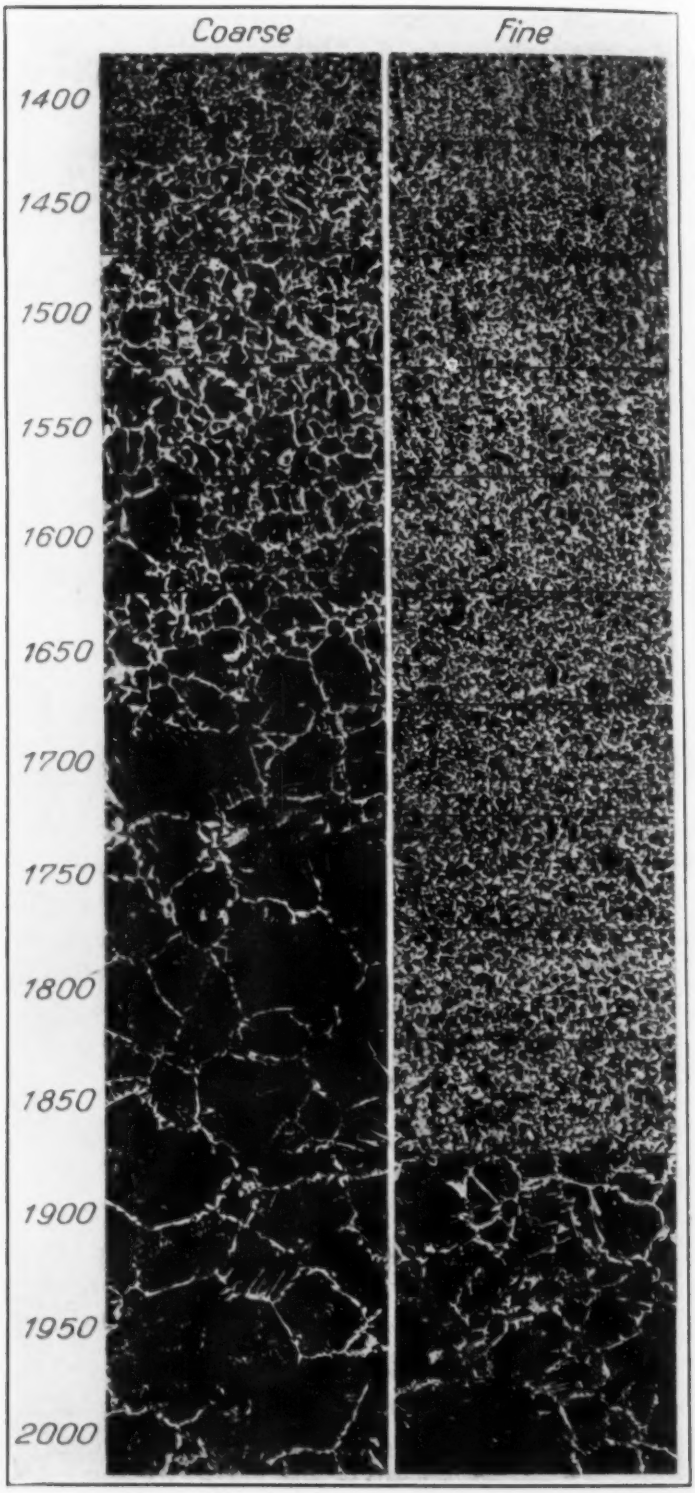


Fig. 3—Composite Micrographs Showing the Actual Grain-Sizes of the Coarse- and Fine-Grained Types After Normalizing at Temperatures Shown.

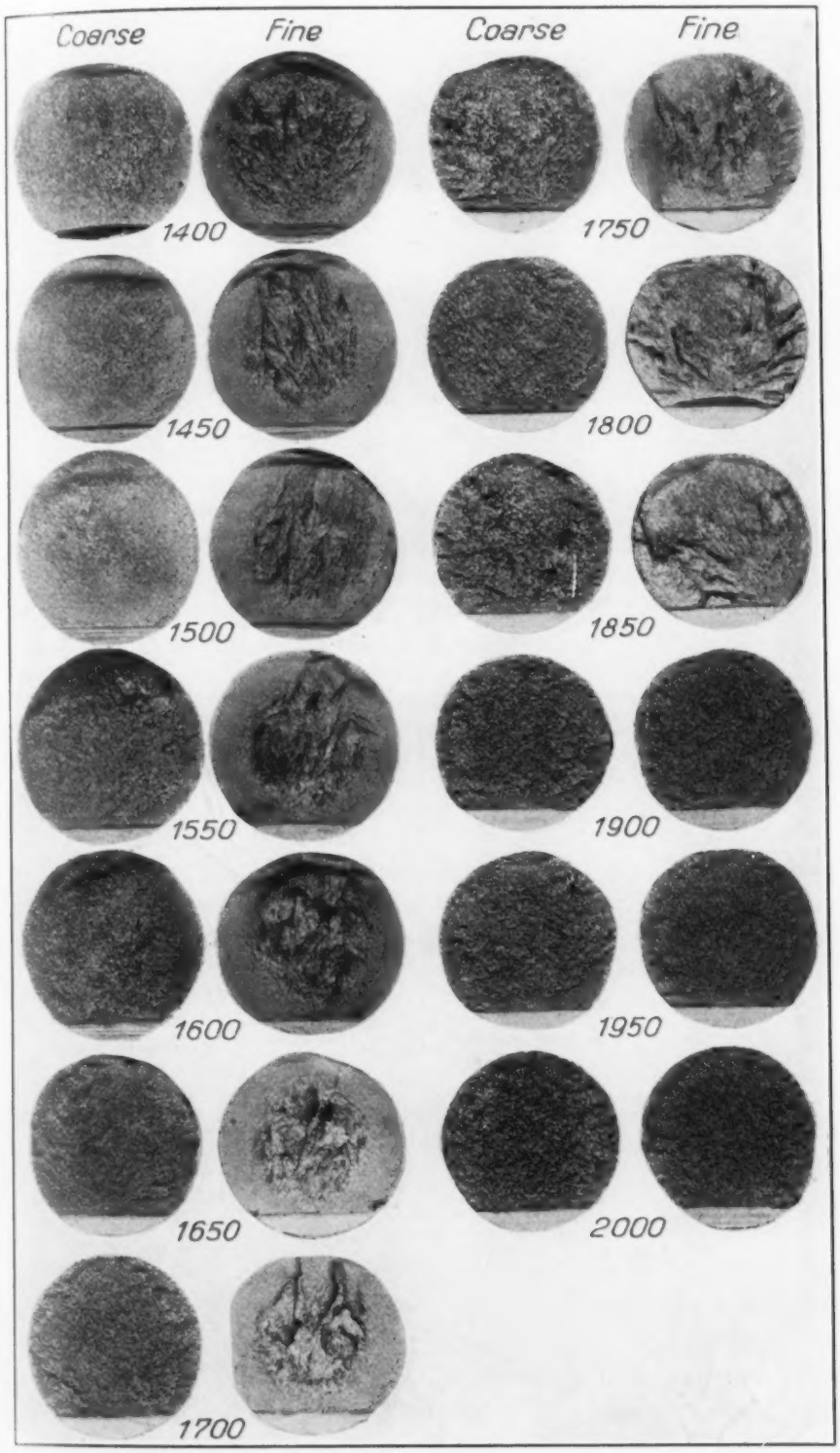


Fig. 4—Fractures of 1-Inch Rounds of Coarse- and Fine-Grained Types, of the Same Heats as Figs. 2 and 3 Quenched as Shown.

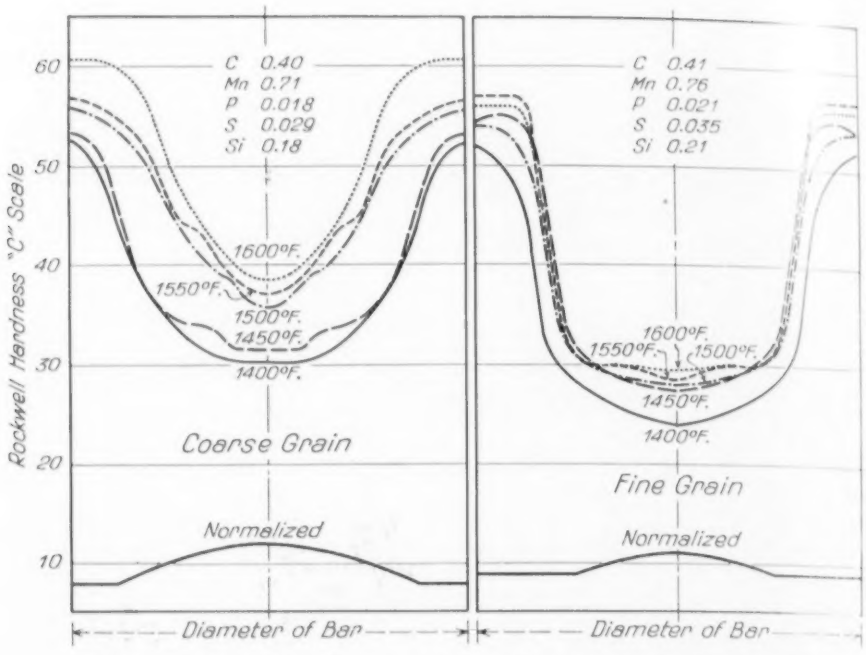


Fig. 5—Hardenability Curves on 1-Inch Rounds of S.A.E. 1040 Coarse and Fine Grain Steels.

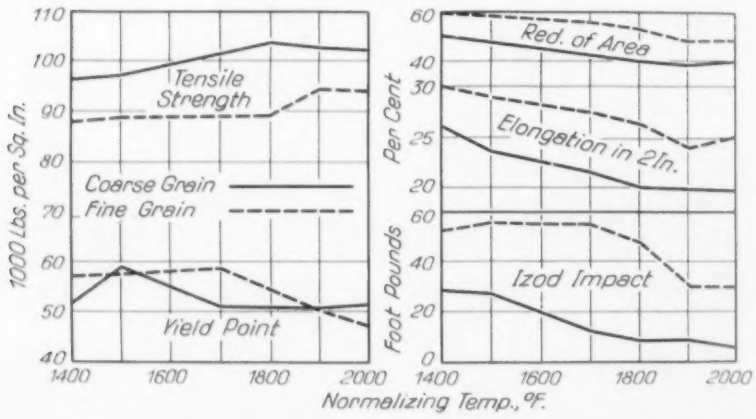


Fig. 6—Comparative Physical Properties of Coarse and Fine Grain S.A.E. 1040 Steel.

types of S.A.E. 1040 steel after quenching in water at 70 degrees Fahr. from 1400 degrees Fahr. to 1600 degrees Fahr. (760-870 degrees Cent.) by 50 degrees Fahr. increments and drawing at 1050 degrees Fahr. (565 degrees Cent.). Treatment was done on 1-inch diameter rounds from which standard 0.505-inch test specimens were machined. These curves show the increased ductility and much higher impact values of the fine-grained type.

Table I

Comparative Physical Properties of Coarse and Fine Grain S.A.E. 1040 Steel

Standard Test Specimens Cut from Center of 1-inch Rounds Normalized for 1 Hour at Temperatures Indicated

Normalizing Temperature Degrees Fahr.	Grain Size	Yield Point lbs./sq. in.	Yield Strength lbs./sq. in.	Tensile Strength lbs./sq. in.	Elongation in 2" %	Reduction of Area %	Izod Impact Ft.-Lbs.
1400	Coarse	51220	53000	95970	26.0	52.2	28
	Fine	56940	55000	87770	30.0	59.3	69
1500	Coarse	58680	56000	96970	23.5	48.0	27
	Fine	57190	54000	88760	29.0	58.0	56
1600	Coarse	55000	52000	99000	22.0	45.0	15
	Fine	58000	52000	89000	28.0	57.0	55
1700	Coarse	50720	53000	101700	21.5	43.6	13
	Fine	58680	54000	88760	27.5	56.2	55
1800	Coarse	50720	52000	103400	20.0	40.6	9
	Fine	54450	54000	89000	27.0	52.7	48
1900	Coarse	50720	52000	102700	20.0	39.4	9
	Fine	50220	52000	94480	24.0	48.5	31
2000	Coarse	51420	53000	102100	20.0	39.7	6
	Fine	47140	50000	93980	25.0	49.4	30

Fig. 9 shows the physical properties of a coarse grain and two fine grain types of S.A.E. 1040 after heat treatment. One of the fine-grained heats, type C, is lower in carbon and higher in manganese than the other types. The surface hardness of the fine-grained steels is approximately the same as the coarse-grained type. These curves also show the increased toughness of the fine-grained types.

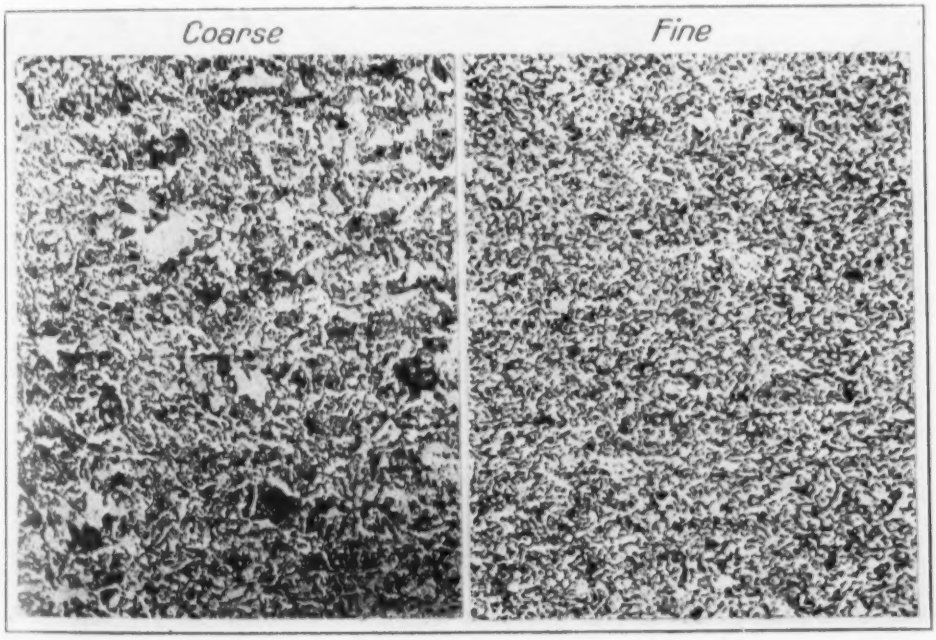


Fig. 7—Micrographs of Fine and Coarse Grain Types of S.A.E. 1040 Steel After Annealing. (× 100)

Table II
Physical Properties of Fine- and Coarse-Grained Types of S.A.E. 1040 Steel After Annealing

Grain-Size	Yield Point	Tensile Strength	Elongation 2"	Elongation 8"	Red. of Area	Brinell Hardness Surface	Brinell Hardness Center	Izod Impact Ft.-Lbs.	Test Specimen
Coarse	44,430	77,140	34.0%	.....	58.1%	...	...	22-19	Machined
Coarse	43,930	75,500	36.0%	21.0%	51.7%	159	149	....	Full Size
Fine	45,430	74,890	32.0%	.....	61.4%	...	...	43.34	Machined
Fine	44,930	73,900	40.0%	24.0%	52.0%	152	143	....	Full Size

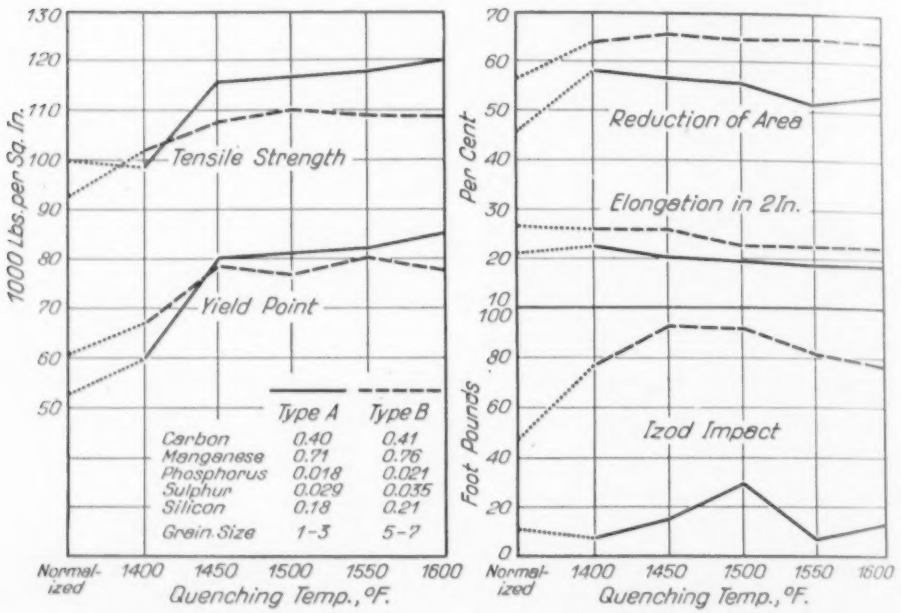


Fig. 8—Comparative Physical Properties of Fine and Coarse Grain S.A.E. 1040 Steel.

Table III
Physical Properties of 1-inch Rounds

Normalized at 1600 degrees Fahr., Quenched in Water from Temperatures Shown, Drawn at 1050 degrees Fahr.

Treatment		Yield Point Lb./sq. in.	Tensile Strength Lb./sq. in.	Elong. 8" -%	Red. of Area	Izod Impact Test Ft.-Lbs.
Coarse Grain Steel (0.40 C., 0.71 Mn., 0.018 P., 0.029 S., 0.18 Si.)						
Water Quenched	1400°F.	59,670	98,210	22.5	58.0	7 - 9
Water Quenched	1450°F.	80,310	115,620	20.5	56.7	12 - 20
Water Quenched	1500°F.	81,310	116,870	20.0	55.9	23 - 26
Water Quenched	1550°F.	82,130	117,580	19.0	51.1	7 - 7
Water Quenched	1600°F.	85,280	119,600	19.0	53.3	11 - 17
Fine Grain Steel (0.41 C., 0.76 Mn., 0.021 P., 0.035 S., 0.21 Si.)						
Water Quenched	1400°F.	66,900	101,850	26.0	63.8	65 - 89
Water Quenched	1450°F.	78,880	107,340	26.0	65.4	90 - 95
Water Quenched	1500°F.	76,890	109,830	23.0	64.5	90 - 92
Water Quenched	1550°F.	80,350	108,890	23.0	64.8	79 - 85
Water Quenched	1600°F.	77,390	108,850	23.0	64.0	76 - 79

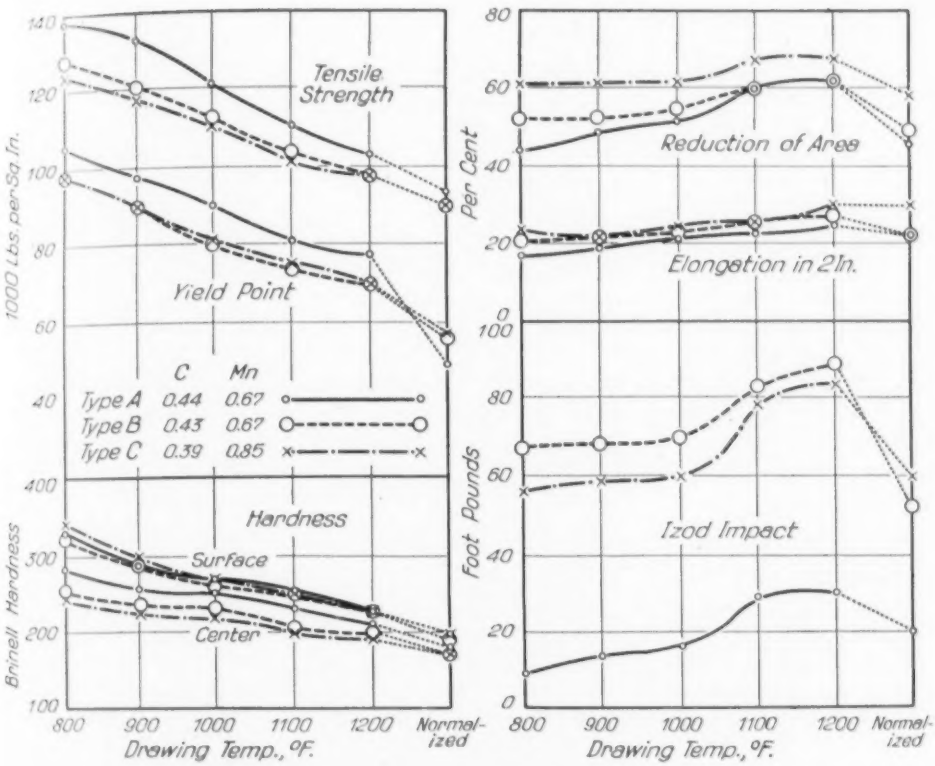


Fig. 9—Physical Properties of a Coarse Grain and Two Fine Grain Types of S.A.E. 1040 After Heat Treatment.

Fig. 10 shows the results of an investigation to determine the cause of "dark spots" in tensile test fractures. It was shown that this condition and the accompanying low ductility are the results of the grain-size of the steel in the "as-rolled" condition, which is controlled by the finishing temperature of the steel as shown by Fig. 2. The finer grain necessary to eliminate these "dark spots" can be more easily obtained by the use of a fine-grained type. This also showed the necessity of control of the McQuaid-Ehn grain-size and the finishing temperatures of steels to be used in the "as-rolled" or "as-forged" condition, in order to obtain the most desirable properties.

Referring to Figs. 6 and 8 and Tables I and III, it will be noted that where abrupt changes of properties are shown in the normalized and quenched and drawn coarse- and fine-grained S.A.E. 1040 steels, they are usually at the temperatures at which the grain structure is changing or where a mixed grain structure is present. This shows very decidedly the influence of grain structure on the

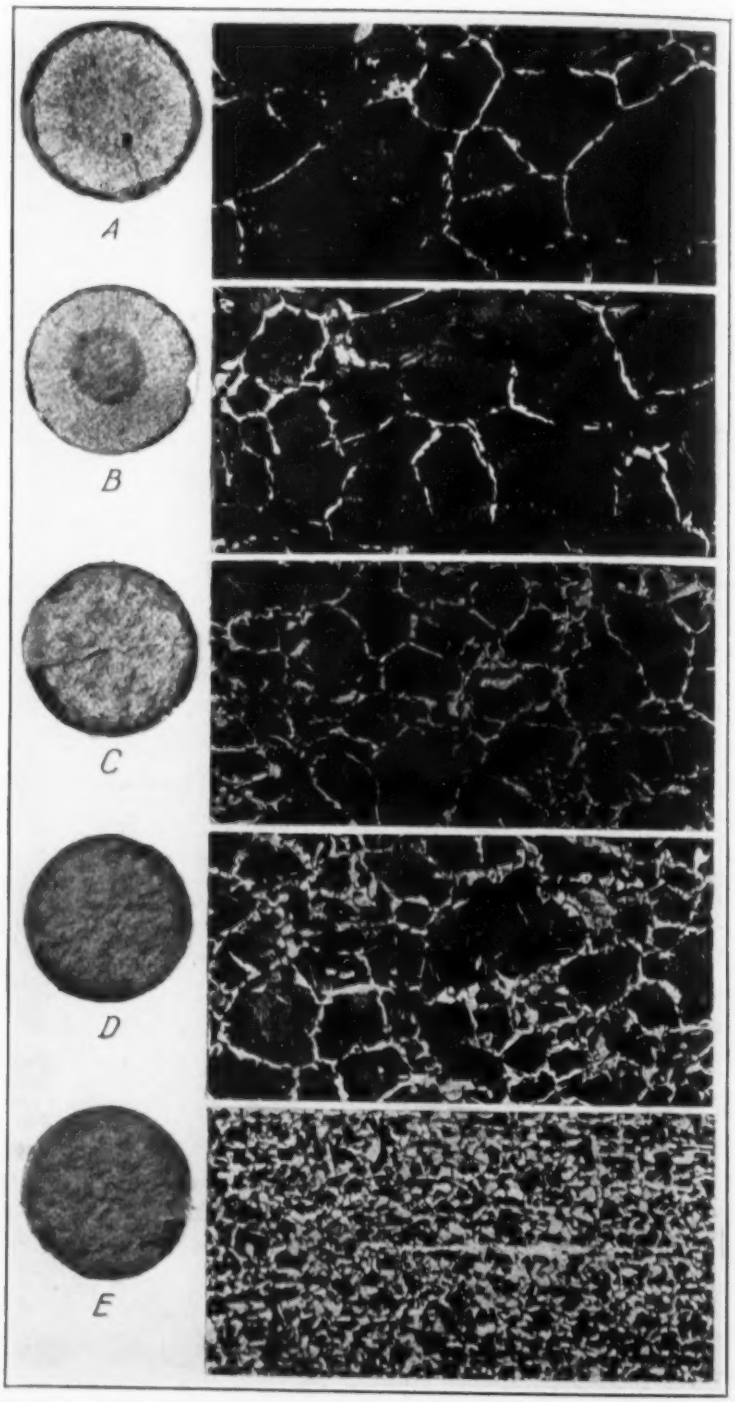


Fig. 10—Results of an Investigation of "Dark Spots" in Tensile Tests Showing the Effects of Finishing Temperatures and Resulting Grain Size. (See Table on Following Page For Physical Properties).

Physical Properties of Specimens Shown in Fig. 10

Specimen	A	B	C	D	E
Approximate Bar Mill Finishing Temperature	2040°F	1980°F	1940°F	1880°F	1775°F
Physical Properties					
Yield Point	56,780	54,600	54,700	56,790	59,900
Tensile Strength	108,500	106,300	102,000	99,190	96,400
Elongation in 2"	24.0%	23.0%	24.0%	30.0%	32.0%
Elongation in 8"	17.5%	15.0%	15.8%	17.8%	21.0%
Reduction of Area	24.2%	26.0%	36.0%	39.2%	41.3%
Fracture	Dark Spots	Dark Spots	½ Cup	Cup	Cup

physical properties and the desirability of a uniform structure at the heat treating temperatures, which can only be obtained by a control of the austenitic grain-size in steel manufacture. Although we have no brief to present for the fine-grained type, it is evident that, in the ordinary heat treating ranges, the hardenability and grain structure of this type are uniform, whereas to obtain this same condition in the coarser grained types it is necessary to go to much higher temperatures which are usually impractical (on account of scaling, decarburization, etc.) and produce less desirable properties. It would seem that the optimum properties for most purposes are obtained by the use of the fine-grained type in the composition necessary to give the desired hardness characteristics and properties.

The data submitted here should demonstrate to anyone associated with the use of steel the impossibility of working with standard heat treating practices to close ranges of physical properties unless grain-size tendencies are controlled. Much of the trouble in mass production operations has been due to variations in the grain growth tendencies of different heats, different parts of the same heat or even in the same forging or sample.

Up to the time when the effects of grain-size were known and controlled steels were available, the natural reaction of the steel user to the unexplainable variations in steel due to lack of grain-size control, was to further restrict ranges of chemical composition, which, in many instances, were already too narrow. It is now possible to obtain improved and more uniform physical properties even when wider ranges of composition and heat treating temperatures are employed.

Also, with this more thorough knowledge, it has been realized that many of the special properties and the uniformity of some alloy steels are not attributable to the alloys themselves, but to the grain-size and uniformity obtained by the use of these alloys and by the closer control exercised over the manufacturing practices

in producing them. With this grain-size control and uniformity possible in carbon steels, new fields have been opened to their use where only alloy steels were formerly considered satisfactory.

Another feature of controlled quality steels is the possibility of using practices which were formerly considered detrimental for steels of certain compositions. For example, higher carbon and manganese steels of controlled types can be successfully quenched in water in sections which previously required quenching in oil to prevent cracking. In most cases the more rapid quench is effective in obtaining more desirable physical properties.

It was difficult to convince some users that fine-grained steels could be successfully case carburized to secure uniform hardness after heat treatment. It has been the opinion that nonuniformity of hardness was due to the abnormality which is usually considered inseparable from fine-grained carbon steels. However, most of the skeptical have been convinced that not only can satisfactory carburizing results be obtained, but the finished part made of fine-grained steel will in most cases possess superior physical properties with less complex heat treatments. For example, higher carburizing temperatures are possible on account of the higher temperatures at which grain coarsening occurs, thereby obtaining the desired depth of case in shorter time. Also quenching directly from the box can be substituted for double treatments in many instances, and better case and core properties obtained. Besides, warpage, which is an important consideration in case hardened parts, is minimized.

With the control of grain-size and composition the steel manufacturer has at hand a means of meeting the specific requirements of almost any particular tool or part and can control the depth of hardness penetration within limits that would be impossible otherwise. If the most suitable steel is once developed for a particular application, a proof test can be devised which will insure future shipments of steel fully meeting the requirements, without the uncertainty that formerly existed when only composition and other less comprehensive standards were used to determine the suitability of the material. The details necessary for the manufacture of the steel are known and their control eliminates uncertainties in the steel making process and makes possible the desired uniformity of product.

THE EFFECT OF McQUAID-EHN GRAIN-SIZE OF STEEL IN FORGING

By W. E. Sanders

Abstract

The scope of this paper is to deal with the improved and more uniform machinability accomplshed by the greater working or compressing of the steel in the processing of the forging; also, the study of the various grain characteristics in steel effecting the density in the bar and the developing of greater density in the forging.

Density, as termed in this paper, is the elimination of the spongy appearance on the surface of a piece, when cut in two, polished and deep etched. By its elimination, tests are shown evaluating the difference in machinability; also, durability is improved by the reduction in gear tooth growth and distortion when hardened, which makes for better alignment and tooth surface contact. By virtue of the better alignment and denser material, pitting and granulations are eliminated, giving greater efficiency and longer life.

THIS paper deals with the improved machinability and durability obtained in automobile transmission gears by so carrying out the process of forging, that the forging is subjected to the greatest possible amount of work or compression; and also with the effect of McQuaid-Ehn grain-size on the development of greater density in forgings. The result of more work or compression on the forging is referred to as density.

Porosity in Deep-Etched Forgings

Nearly every one having to deal with steels and forgings has at sometime or other cut pieces in two, polished the faces and etched them, and after etching has found a number of tiny cavities or voids (sometimes called spongy metal) occurring at various places, sometimes more prominent at the center or core and at other times near the surface. The location, nature, and extent of these voids or cavities, as brought out by the etch test, determines the degree of

A paper presented as part of the Grain-Size Symposium of the Sixteenth Annual Convention of the Society held in New York City the week of October 1, 1934. The author, W. E. Sanders, a member of the Society, is metallurgical engineer, Delco Products Division, General Motors Co., Dayton, Ohio. Manuscript received July 5, 1934.

density of any forging. The author's theory is, substantially, that in forging it is possible to produce internal slipping or tearing of the fibers in the steel. This slipping, or tearing, is indicated by the voids or cavities brought out by acid etching. Some previous papers on the subject attribute the general porosity in etched forged sections entirely to the quality of the steel, this quality being determined by the amount and nature of the nonmetallic substances present. With this thesis the author cannot entirely agree, although we recognize that sulphur and other impurities existing in the steel to too great an extent will make the grain-size appear coarse or open. In addition to this, however, the author is endeavoring to bring out the fact that the largest percentage of openness or sponginess is due to torn fibers; also, that there are some steels of the same analysis containing approximately the same amount of nonmetallic impurities, which after proper forging do not show this porosity.

Having observed a large number of etched forgings in which the porosity or density was almost negligible in some, and quite pronounced in others, tests were made in which these conditions were found to be associated with the variables in machinability in the machine shop.

Eliminations of Tearing of the Fibers in Steel During Forging

Density, as we term it, results from the elimination of the tearing of the fibers during upsetting or forging. In working the steel, we must first bear in mind that in reducing or upsetting the metal too much in one pass, the fibers will be broken, as in a tensile test, before the internal structure can adjust itself to the movement. Also, in the working of the steel, flow takes place within the grains, and there must, be movement at the boundaries, accompanying the internal movement of the grain. This integranular movement must be greater or less, according to the irregularities or nonuniformity of the internal flow. When these irregularities are great, ruptures will occur either at the boundaries or through the grain, and the material will show voids or cavities upon etching.

Gough[1] in his work, on "Crystalline Structure in Relation to

[1]The 1933 Edgar Marburg Lecture sponsored by the American Society for Testing Materials and the Engineering Section of the American Association for the Advancement of Science. The lecturer was Herbert John Gough, F.R.S., superintendent of engineering department of the National Physical Laboratory, Teddington, England. His subject was "Crystalline Structure in Relation to Failure of Metals—Especially by Fatigue."

Failure of Metals—Especially by Fatigue," makes the following statement: "Much greater interest attaches to the very accurate determinations of the changes in density which took place during the fatigue stressing. The density of the material in the form of the *aggregate* (*many grains*) definitely decreased by amounts varying from 0.037 to 0.100 per cent of the original value, while that of the *single crystals* remained unchanged. These results, which are in agreement with results obtained by other workers using static stressing, suggest that the decrease in density of metals due to cold work marks an effect which is confined to the neighborhood of crystal boundaries where no such boundaries are present, as in single crystals, changes of density do not occur."

Forging Annealed Bars to Increase the Ductility of the Fibers

With these theories in mind, our attention was drawn to the study of the steel prior to forging for its characteristics that might produce more ductile fibers, thus promoting flow of the metal without tearing.

This suggested experiments having as their object a coarsening of the structure of the forging bars prior to heating for forging. It was decided to accomplish this by annealing at high temperatures. A bar of G.M. 5150-A* steel was selected which had a medium-fine structure in the rolled condition. It was then cut into twelve lengths suitable for making the same kind of forgings. Four of the pieces were forged as received and the balance annealed by furnace cooling after soaking for one hour at 1650 degrees Fahr. (900 degree Cent.). Half of these forging lengths were then upset into forgings like those which had been produced from the original bar. The remaining pieces were then reannealed by furnace cooling from 1750 degrees Fahr. (955 degrees Cent.) after soaking one hour, and were then upset into the same kind of forgings. Microscopic specimens, prepared from the bars in each group before forging, showed the structures to have been coarsened considerably after each anneal.

The three groups of the forgings were then cut and etched. Typical examples of each group are shown in Fig. 1, A representing the structure of the forging made from the untreated bar; B, after the first anneal; and C, after the double anneal. It will be observed

*G.M. 5150-A Steel is similar to S.A.E. 5150 Steel. The composition range is

C	Mn	P (max.)	S (max.)	Si	Cr
0.48-0.53	0.60-0.90	0.040	0.050	0.15-0.30	0.80-1.10

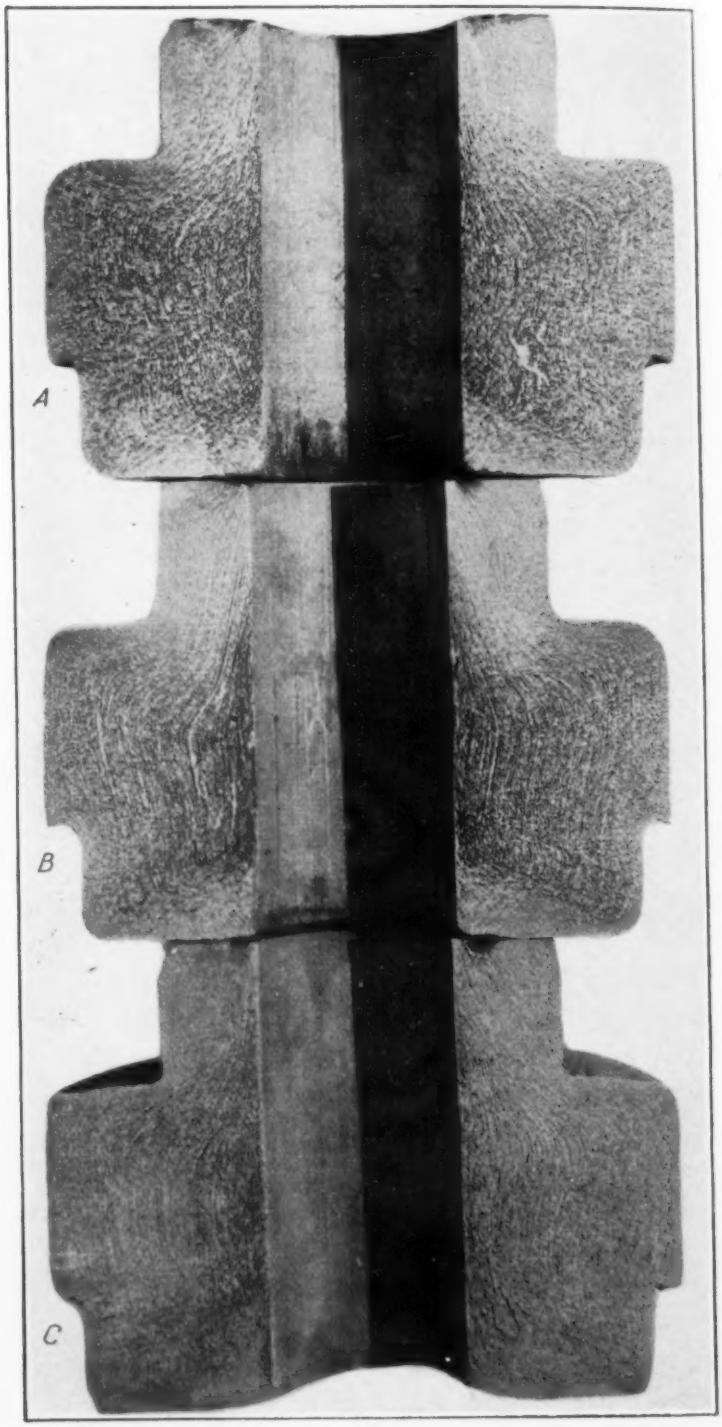


Fig. 1—Sectionalized Forgings Deep Etched.
A—Structure As-Forged from Untreated Bar.
B—Structure As-Forged After First Anneal.
C—Structure As-Forged After Double Anneal.

from this figure that there is a progressive decrease in the amount of torn fibers from A through to C, and a like increase in the continuity of the flow lines. The increased density of forgings B and C made it necessary to etch them ten minutes longer to bring out the flow lines. This experiment is significant in that it suggested that a coarse structure has more ductile fibers, leading to further study of the steels which are inherently coarse-grained.

Characteristics of Inherently Coarse and Fine Grain Steels

We must remember that all steels have what is known as 'fiber' to a greater or less degree. A steel made fine-grained, according to the McQuaid-Ehn test, is usually more fibrous, or rather the fibers are finer and shorter, than a steel that is made inherently coarse-grained, assuming that both steels are poured into the same size ingot. The fine-grained steel, when given the same heat treatment as the coarse-grained steel, will show approximately the same values for tensile strength and elastic limit, but decidedly superior ductility values in the static test piece taken longitudinally to the direction of rolling. The dynamic test piece machined longitudinally to the direction of rolling and broken across the fiber or grain will show decidedly superior values under impact for the steel made inherently fine-grained. When, however, a static test is made transversely to the direction of rolling, the tensile strength and elastic limit are again approximately the same, but the ductility values as shown by the elongation and reduction of area are now decidedly greater in the case of the steel made inherently coarse-grained. The transverse impact piece machined transversely and broken parallel to the direction of rolling will show much better resistance to shock in the case of the coarse-grained steel. In the upsetting of gear forgings, the steel is subjected to severe transverse stress.

The work of Charpy and others has shown that excessive reduction from the ingot to the bar lowers the transverse test values of steel, especially the dynamic test values. One of the reasons that a crucible-melted steel performs better than a similar grade of steel cast in a larger ingot is that with a smaller ingot a more uniform and quicker freezing occurs with smaller dendritic growth and a more uniform distribution of carbon, etc. The smaller the ingot, the fewer and also shorter will be the number of fibers in a bar rolled from that ingot. Also, the larger the ingot, the more is the likelihood of the nonmetallic particles, carbides, etc., being thrown

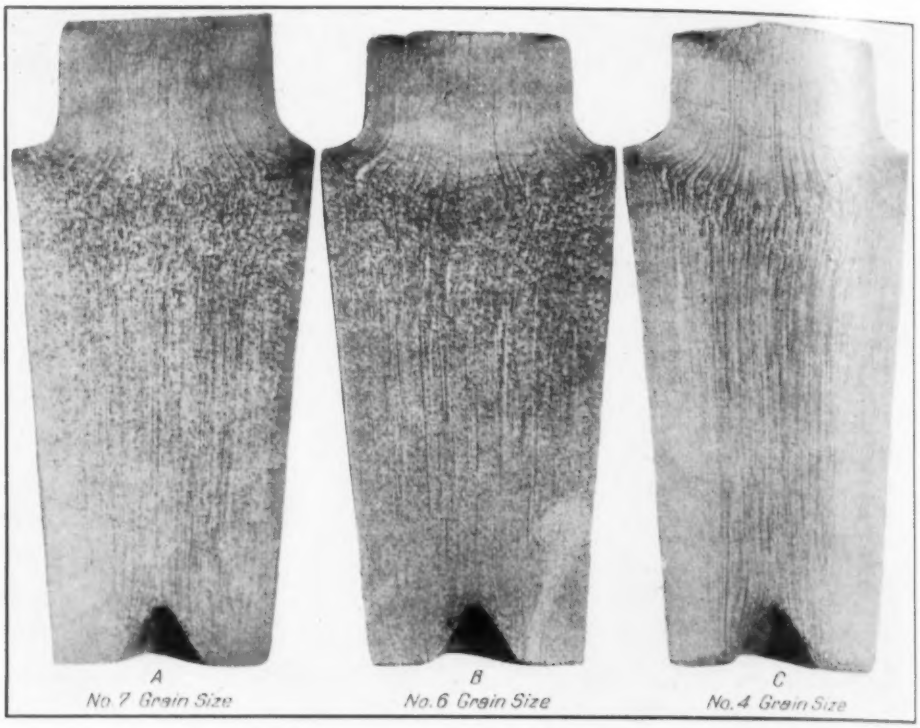


Fig. 2—McQuaid-Ehn Ratings of Gathmann Type Ingots.

into the center before freezing occurs. It seems, therefore, that the quicker freezing occurs in the ingot, the more uniform will be the fiber condition from skin to center of the bar and the more uniform will be the density of the bar.

Forging as Related to McQuaid-Ehn Inherent Grain-Sizes in Steel

With this knowledge at hand, arrangements were made with the steel companies to furnish a few bars of different steels, having inherent grain-size ratings varying from 4 to 9. The heats for producing these bars were poured into various size molds, ranging in ingot sizes from 12 to 28 inches, all molds being of the Gathmann type.

These bars were rolled to the same diameter and upset into disks in ratios to the bar—from 1½ to 1, in one-half inch steps, up to 7 to 1. It was noted in these tests, which were made from G.M. 5150-A steel, that the coarse-grained steels, by virtue of greater ductility and longer fibers, showed less tearing in the upsetting. More tearing, less continuity of the fibers or more porosity was obtained

in the upsetting of the finer-grained steels. In these experiments the number 4 grain, molded into a 19 inch ingot of the Gathmann type, showed the greatest amount of density when etched.

Bars from this ingot size having McQuaid-Ehn ratings of 4-6 and 7 were used for upsetting a cone on the end of the bar. These ingots of the same analysis and upset at the same temperature, the McQuaid-Ehn rating being the only variable, are shown in Fig. 2. It will be noted that A is very coarse, or open and was upset from McQuaid-Ehn number 7, while B is less coarse, or open, upset from number 6, while C upset from number 4 shows evidence of more continuity of the fibers, affording greater density.

Counter gears with pierced holes were also upset from this same group, with the repetition of the results shown in the disk and cone experiments. In this experiment a marked difference was also noted in the piercing, in favor of the coarse grain steel. The ductility of the fibers of the coarse-grained steel not only makes it possible to obtain greater density of the metal in upsetting, but also permits the piercing of smaller holes of greater depths or lengths.

Recognizing that the development up until this time on dense homogeneous structures had continually resulted in increased machinability and durability, while specimens continued to show some minute cavities or voids upon etching, it was felt that some further study should be made to see if these voids could be entirely eliminated, and if so, whether still greater machinability could be attained.

High Temperature Diffusion Experiments

For the reasons just stated, attention was called to the work of Ishiwara[2] on diffusion of metals at high temperatures, which suggested a research on the forging of diffused materials. In these experiments, eighteen cycles were run at different temperatures between 2370 and 2500 degrees Fahr. (1300-1370 degrees Cent). The bars after being cut to forging lengths were sealed into tubes of heat resisting material by welding a disk over each end of the tube; this was done to prevent a reduction of the bar size by oxidation. The tubes were then put into a small furnace which was automatically controlled and heated to the different temperatures, and held at these temperatures from one to three hours and cooled at different rates. One of the experiments carried a cycle of heating

[2]Torajero Ishiwara, "The Effect of Impurities on the Dendritic Structure in Carbon Steels, and their Diffusion at High Temperatures," *Science Reports*, Tohoku Imperial University, Sendai, Japan.

to 1750 degrees Fahr. (955 degrees Cent.), holding for three hours, after which it was cooled to 1200 degrees Fahr. (650 degrees Cent.) at the rate of 100 degrees Fahr. per hour, reheated to 2370 degrees Fahr. (1300 degrees Cent.), held for three hours, and cooled to 1200 degrees Fahr. (650 degrees Cent.), at the rate of 100 degrees Fahr. per hour and discharged from the furnace for further cooling in air. This cycle was significant in that the McQuaid-Ehn grain-size had changed (in addition to the structural grain) from originally 4-6 to a definite 3. It remained unchanged after normalizing the second time at 1650 degrees Fahr. (900 degrees Cent.). This cycle was repeated, with the result that the grain-size had increased the same as before, and after the sixth normalizing, the grain-size still remained coarse; and even in the third and fourth normalizing the grain-size tended to coarsen even more, and remained so until the fifth and sixth normalizing where it became a point or two finer, but never returned to its original size.

In the cycle just explained, it was found that the change in the inherent grain-size occurred when cooled from 1950 degrees Fahr. (1065 degrees Cent.), and if the heating was continued to beyond 2000 degrees Fahr. (1095 degrees Cent.) before cooling no change was effected. Therefore, from this experiment, in this particular steel, there seems to be a transformation point in the inherent grain, when heated and cooled in the range between 1950 and 2000 degrees Fahr. (1065-1095 degrees Cent.).

Other cycles employed, increased the size of the structural grain, with no effect upon the inherent grain. From each of these cycles, the bars were forged into counter gears, which were cut in half and etched. All of the forgings from these heats displayed greater density. The density of the forgings was greater when made from the bars displaying increased growth of the inherent grain.

Billets were also diffused at the mill, some of which were rolled into 1⅞ inch round bars, which later were upset into 5 inch diameter disks 1 inch thick. Specimens were taken from the billet, bar and forgings, before and after diffusing, for physical tests, the results of which are shown in Table I.

Forging Process and Its Effects

We must also bear in mind that in addition to the effects of grain-size in forgings[3], there are many variables in the actual forg-

[3]The author's paper on "Researches on Steels and Forgings for Greater Density, Machinability, and Durability," given before the American Society of Mechanical Engineers, Ontario, Canada, June 29th, 1932.

Table I

Comparison of Mechanical Properties Before and After a Diffusion Heat Treatment
S.A.E. 5140 Steel

	Yield	Ultimate	Elong. 2"	R.A.	Izod	Brinell Tensile	Brinell Izod
Diffused 24 Hrs. 6"x6" Billet							
Transverse	120,000	136,000	14.0	40.4	29.0-27.5	285	269
Longitudinal	113,000	130,500	20.0	59.1	63.5-65.0	277	269
Diffused 3 Hrs. 6"x6" Billet							
Transverse	117,000	133,500	15.0	37.9	28.0-27.0	277	269
Longitudinal	115,000	132,000	20.5	62.1	65.0-63.5	285	269
Not Diffused 6"x6" Billet							
Transverse	117,000	133,500	15.0	35.7	24.0-24.5	277	269
Longitudinal	117,000	133,500	20.5	62.1	69.5-65.5	277	269
Diffused 24 Hrs. and Rolled to 1⅞" Round							
Longitudinal	123,500	139,000	22.0	60.8	62.5-59.0	285	285

1⅞" Round Upset Series
7-Inch Lengths of 1⅞-Inch Round Upset to 5-Inch Diameter Disks 1 Inch Thick

No.	Yield	Ultimate	Elongation	Reduction Area	Izod
Upset Bars—Diffused Afterwards					
1	113,000	130,500	18.5	57.3	40.5-44.0
2	113,000	131,000	18.5	56.8	45.0-40.0
Diffused Bars—Upset Afterwards					
3	113,000	131,000	17.0	51.7	40.0-41.5
4	113,500	132,000	18.5	55.2	29.0-31.0
Upset Bars—No Diffusion Treatment					
5	112,000	127,500	17.0	53.6	35.0-37.0
6	113,500	131,000	17.5	51.4	36.0-44.5

ing process which affect density. These alone, if corrected will improve machinability and durability to a great extent in any grain-size specification. They are not difficult to control when specifications are made to cover all of the variables. First, the steel should be purchased on density and grain-size specifications, in addition to the chemical specification. Second, the forging dies must be designed to effect maximum compression on the metal in the forging. Third, the furnaces should be of a design that will insure uniform temperature throughout.

At the beginning of the research, forgings were being made without any temperature control, as low as 1800 degrees Fahr. (980 degrees Cent.) to a maximum of 2000 degrees Fahr. (1095 degrees Cent.). Temperatures were at once increased to 2050 to 2150 degrees Fahr. (1120-1175 degrees Cent.) using optical pyrom-

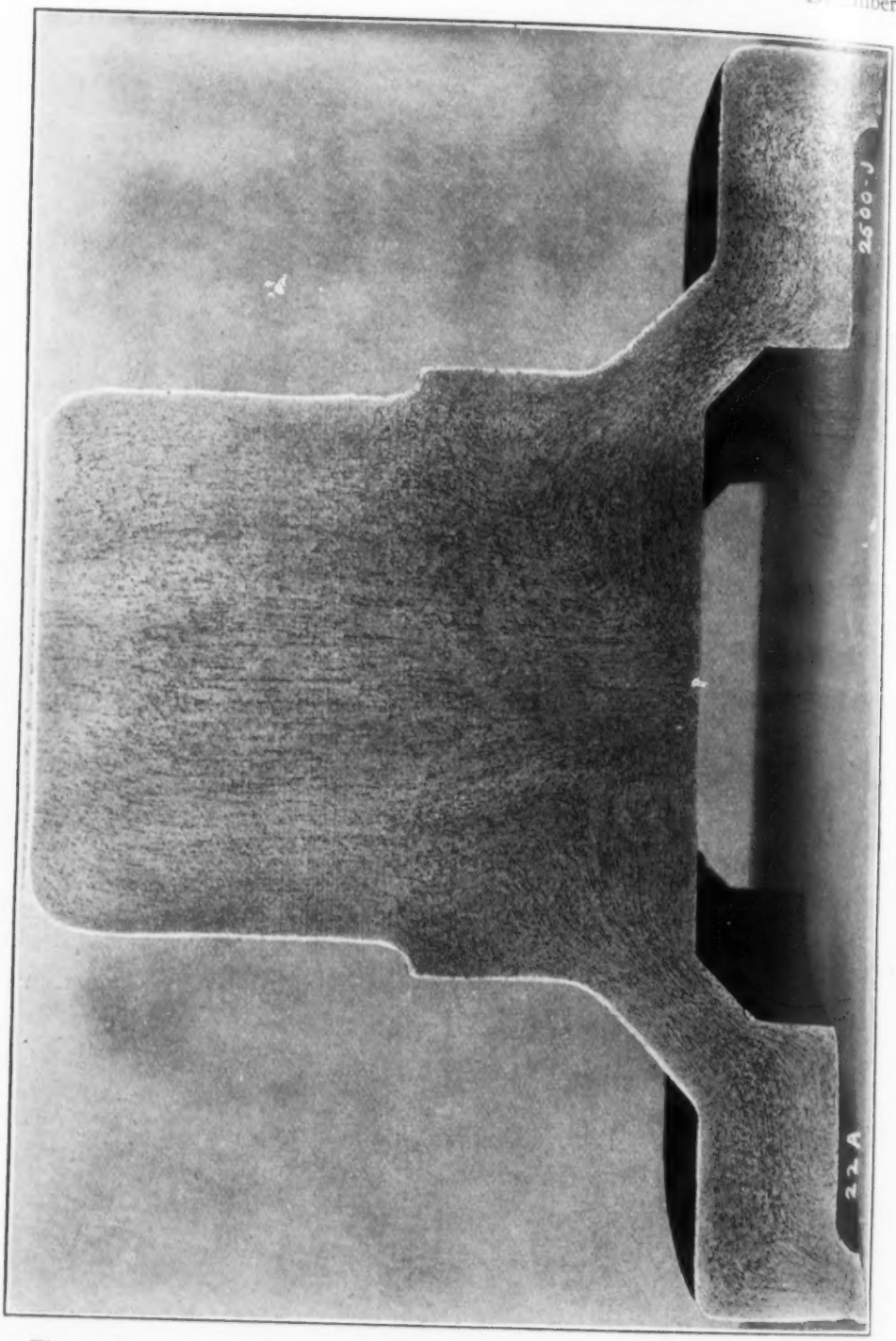


Fig. 3—Structure of a Deep Etched Forging Which Had Good Machining Qualities.

eter for control. Improvement in machinability and structure was noted with this increase in temperature, producing structures comparable to Fig. 3. With the evidence at hand on coarse grain and the results of the high temperature diffusion, research on forgings forged at higher temperatures was suggested.

Test forgings were forged at 2200 and 2300 degrees Fahr. (1205-1260 degrees Cent.), using material having a number 4 grain-size rating. These forgings when cut and etched showed a dense homogeneous structure, and were equivalent to the forgings made from diffused material. From these results, forging temperatures were increased for production from 2200 to 2300 degrees Fahr. (1205-1260 degrees Cent.) and machining performance was comparable to that obtained on diffused material. The evidence at hand up to this point was developed on open-hearth steels.

Machinability Tests

One of the tests, of which quite a number have been made, is one shown in the photomacrograph Fig. 4, No. 1, and Fig. 5, No. 2. This test was made on pieces taken from a dense structure and one not quite as dense as shown by the deep etch, to evaluate the difference in machinability. These pieces were taken from two clutch gears (at locations shown in photomacrographs) and prepared by being cut and ground into ½-inch square cubes. They were taken from the longitudinal and transverse sections of the stem or original bar and from a section in the forging. The test was run by placing a thermocouple in the piece to register the heat generated at the point of the drill by drilling to a uniform depth with a uniform pressure on the drill in each piece. The percentage of difference in machinability is shown by the milli-volts generated in the thermocouple.

Milli-volt Readings Taken in Forging	*No. 1*	*and No. 2*
Longitudinal Sections of the Stem	3.66	3.94
Transverse Section of the Stem	4.04	4.27
Section from the Forging	4.04	4.53
Difference in Per Cent Between Sections		
Longitudinal of Stem and Forging	8.8	13.0
Transverse of Stem and Forging	0	5.6
Longitudinal and Transverse of Stem	8.8	7.8
Difference in Per Cent Between Forging	*No. 1*	*and No. 2*
Longitudinal of the Stem	7.0	
Transverse of the Stem	5.6	
Section in the Forging	11.0	

Another test on dense and porous forgings, similar to the forgings shown in Fig. 4 and 5, except that the difference between

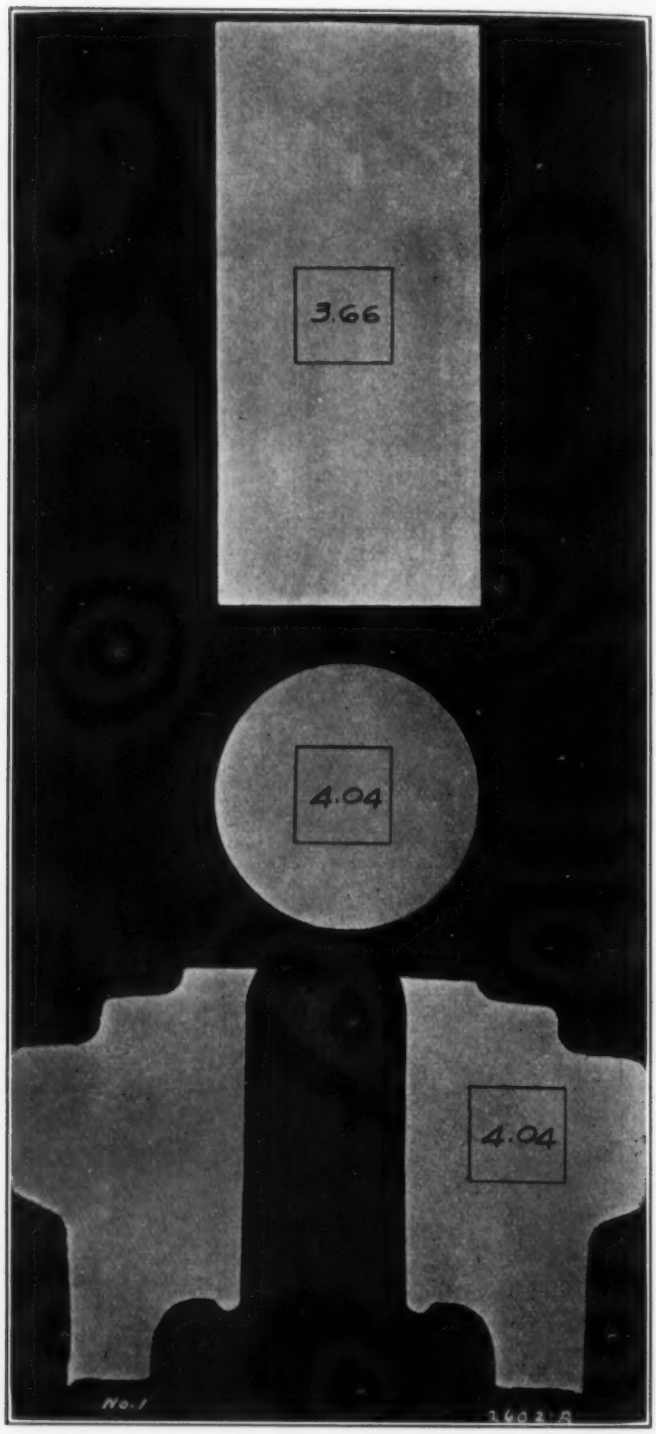


Fig. 4—Sectionalized Dense Structure Forging Used in Machinability Tests.

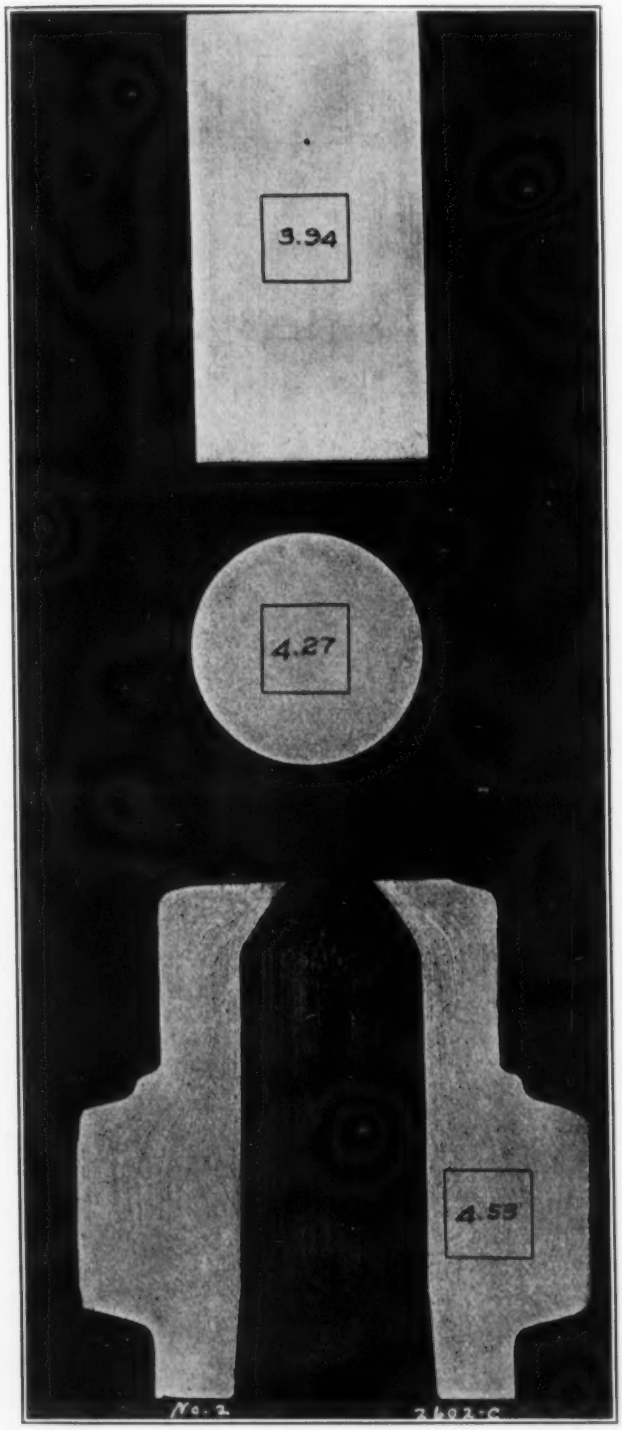


Fig. 5—Sectionalized Less Dense Structure Forging Used in Machinability Tests.

the dense and porous was greater as indicated by the etch, and also by a difference of 22 per cent in the thermal measuring test, were taken to the University of Michigan for further check. They were put on a drill dynamometer indicating torsion and pressure at a constant feed.

In drilling the porous forging the torsional indicator hand vibrated considerably (from 7 to 10 points), and showed 8.3 per cent greater torsion than shown when drilling the dense forging. When drilling the dense forging the torsional indicator hand was very steady, only vibrating from one to two points, with the pressure indicator showing approximately the same in both forgings.

A similar test, which more or less confirms the drill dynamometer, was run on a drillpress with a polyphase graphic wattmeter connected in the motor circuit. The drill was arranged for drilling dense and porous forgings, with the same speeds and constant automatic feed; the results of which are shown in Fig. 6. It will be noted from the chart, the porous material not only required more power, but slowed down the motor to the extent of requiring longer time to complete the operation. The additional time and watts averaging 26.3 per cent in favor of the dense forgings.

From these findings and evaluations, one may appreciate the possible increased machinability in the bar, if selected for dense homogeneous structure. One manufacturer reports increase feeds and speeds with a tool saving of 58 per cent by purchasing selected material.

Open Hearth and Electric Steels

Evaluating the difference between open-hearth and electric furnace steels, let us refer to Figs. 4 and 5. Number one was forged from G.M. 5140-A* electric furnace steel, received and accepted as good dense structure in the bar. Number two was forged from G.M. 5140-A open-hearth steel, received and accepted as a fairly dense structure in the bar. The forging when cut and etched showed a structure density comparable to the bar when properly forged. Bars of good dense structure may be improperly forged (showing porosity when etched) causing a decrease in machinability. Also a bar of poor structure cannot be forged to greater machinability than obtained from the bar, and more often it will not equal the bar. Incidentally, it is these variables, that many times necessitate forgings to be made of higher priced steels.

*G.M. 5140-A Steel is similar to S.A.E. 5140 Steel. The composition range is

C	Mn	P (max.)	S (max.)	Si	Cr
0.38-0.43	0.60-0.90	0.040	0.050	0.15-0.30	0.80-1.10

A certain amount of evidence is at hand which indicates that electric furnace steels will run more consistently dense than open-hearth steels. This to a certain extent automatically produces a dense homogeneous structure when forged. Also, there is a certain amount of evidence which indicates that forgings made from

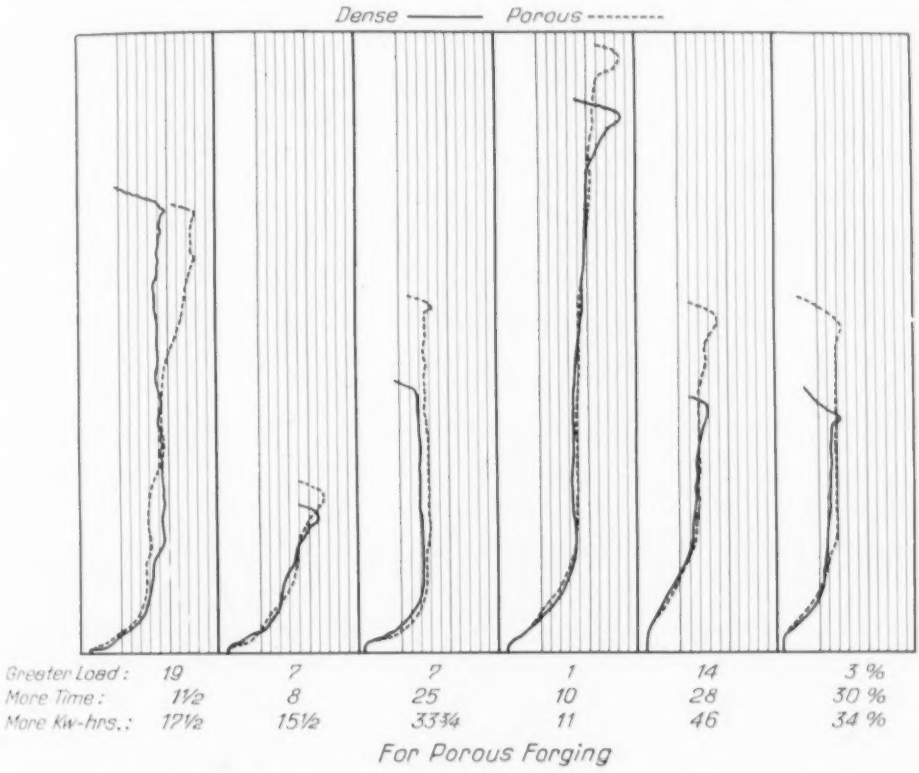


Fig. 6—Tests Made on Porous and Dense Forgings by Means of a Polyphase Wattmeter.

electric furnace steel, comparable to the bar shown in Fig. 4 No. 1, but due to porosity from improper forging, show a decrease in machinability of from 20 to 40 per cent after forging. It may also, be said that Fig. 5 No. 2, in the bar is comparable to the maximum density that consistently may be expected from commercial open-hearth practice. There is also further evidence at hand which indicates that Fig. 5 No.2 forging could have been forged to approximately 5 per cent greater density, or comparable to the bar. Therefore, if the steels are received and forged to obtain maximum density of open-hearth practice, then from this test 5 to 7 per cent greater machinability can be obtained by using electric furnace steel.

There is also a certain amount of evidence, whether made in

the open-hearth or electric furnace, that the coarse-grained steels are fundamentally cleaner with less nonmetallic film material than the finer grain steel and consequently more uniform. This also is one of the fundamental reasons that preference was given the coarse-grained steel for dense homogeneous structure.

Improved Machinability and Distortion

By correcting the forging practice, an improvement in machinability may be expected from 15 to 50 per cent, depending on the operation. Speeds and feeds may be increased from 11 to 60 per cent and tool life increased in like proportion.

To a gear manufacturer there are other things of major importance besides machinability, that is, changes in the heat treatment. If these changes can be controlled and made uniform, the change in cutting can be allowed for and becomes of little consequence.

With a dense homogeneous structure, distortion becomes more uniform in hardening, enabling accuracy of spacing maintained on hobbing machines within 0.0002; contour within $\pm$0.00025, and when the pieces return from the fire you can expect 0.0004 to 0.0005 inch change in contour and no apparent change in spacing. All checks being made with teeth marked and starting from same position, carrying through in the same direction. The gears going to the hardening department with concentricity of less than 0.001 return with 0.0015 inch. With the control maintained on the steel and forgings, it is a simple matter to correct the hob by grinding more or less rake in the cutting face, thus compensating for the regular fire distortion.

Durability is likewise effected by the reduction in tooth growth and distortion, which makes for better alignment and tooth surface contact.

Conclusion

In this development it has been found possible to produce denser forgings, in which slipping and tearing of the fibers are minimized by employing a steel of a specified grain-size and structure, accurate forging temperatures, and specially designed dies for controlling the flash to insure greater compression of the forging.

It is recognized, however, that in carburizing steels, where the carbon is usually low, the inherent grain-size is not so important

and should not be coarser than a number 5 for good case refinement with the use of ordinary treatment.

With the evidence at hand, of the effect of mechanical working being concentrated at the grain boundaries, it is hard to avoid the conclusion that the manner of hot working the steel has something to do with the density of the material at the grain boundaries. To carry the speculation further, the forging of large-grained structures would be preferable because they would contain fewer grain boundaries along which density changes could take place. This interpretation would merely be transferring the same thought from the notion of a uniform change in density throughout the structure to a concentrated change in density occurring only in the region of the crystal boundaries.

With this in mind, the object of a properly controlled forging process would be to cause as little disruption as possible of the more or less stable material at the grain boundaries.

DISCUSSION

Written Discussion: By Frederick G. Sefing, Michigan State College, East Lansing, Michigan.

The writer wishes to draw attention to the author's statement that the impact test bars broken across the fibers are superior when fine-grained and when broken parallel to the fibers the coarse-grained shows superiority. Although gear forgings may well be made with coarse-grained steels, as the author shows, the common run of forgings require fine-grained steel (about No. 6 McQuaid-Ehn) in order that the longitudinal test bars show maximum toughness. It is obvious, therefore, that the author's specifications cannot be applied at present to the latter class of forgings.

The writer has observed, however, that the impact values of forgings do not bear a close relation to the inherent grain-size. Izod bars were prepared from eight different heats of S.A.E. 1045 steel all having a grain-size of 4 (McQuaid-Ehn). The test bars were machined from ¾-inch bars forged from billets and hardened and tempered without normalizing. The impact values varied from 37.75 to 62.75 foot pounds.

The author refers to a uniform compression of the steel during the forging operation, which results in a denser structure and obviously more uniform properties of the forging. However easy this may be in the class of forgings as explained by the author, yet tearing and lack of uniformity is apparent in the upset rim of Fig. 3 of the paper as compared with the large center section. In the shaping of larger pieces and particularly forgings of varying thickness the desirable dense and uniform steel would be a remarkable feat at any temperature in any forge shop. Uniformity of metal structure and of

physical properties of the common forgings, therefore, are a difficult accomplishment in modern shop practice.

The author's paper shows unusually low distortion of the machined forgings during hardening. In a series of experiments by the writer, with S.A.E. 1045 steel, the distortion was observed to be considerably greater than the results presented in the author's paper. Furthermore, inconsistencies in the volume changes were prevalent as a result of heat treating the forged steel. These variations were found to be of the same order of magnitude between samples cut from the same bar as between the different steels. The volume change upon normalizing 2½-inch diameter bars, forged from billets, was from 0.28 per cent expansion to 0.76 per cent shrinkage between steels and from 0.08 per cent shrinkage to 0.76 per cent shrinkage between different samples from the same bar. The change upon hardening and tempering the normalized bars was from 0.07 per cent shrinkage to 0.35 per cent expansion between steels and from 0.07 per cent shrinkage to 0.24 per cent expansion between samples from the same bar. It is to be noted that still greater variations have been observed as a result of hardening and tempering the "as-forged" steel. Thirty-seven heats of S.A.E. 1045 steel gathered from various sources were investigated in the writer's experiments. The grain-size ranged from 3 to 8 (McQuaid-Ehn). In these volume determinations by the writer precautions were taken, of course, to keep the scaling effect to less than $\frac{1}{10}$ of the effect to be measured. It is to be borne in mind that the writer's results were obtained after studying only 37 heats of one grade of steel and that the indications of this work are being confirmed by additional tests.

In the writer's opinion, therefore, the use of the McQuaid-Ehn grain-size of steel as a guide to forging qualities is for the present at least very limited. The writer's results indicate moreover that no grading of forging steels as to normality or grain-size can be used in the present state of the art to control the subsequent volume change.

The author is to be commended for setting before the forging industry a new ideal for producing parts of improved properties, and a means for accomplishing this ideal in one class of forgings.

INFLUENCE OF GRAIN-SIZE ON THE HIGH TEMPERATURE CHARACTERISTICS OF FERROUS AND NONFERROUS ALLOYS

By A. E. White and C. L. Clark

Abstract

This paper gives results which show the influence of both the McQuaid-Ehn (inherent) grain-size and the grain-size as produced through heat treatment (actual) on the high temperature characteristics of metals. For the inherent grain-size studies two ferrous alloys were used, one of which contained 0.50 per cent molybdenum and the other 0.25 per cent molybdenum-1.25 per cent manganese. For the actual grain-size investigations two nonferrous alloys of the copper-zinc-tin type were used which differed in the amounts of copper and zinc which they contained.

The influence of inherent grain-size was determined by tensile and Charpy impact tests at both room and elevated temperatures and by creep tests at 800, 1000 and 1200 degrees Fahr. (425, 540, 650 degrees Cent.). Results are also given from tests conducted on the completed creep specimens for the purpose of determining the structural stability of these alloys under the given test conditions.

The tests on the nonferrous alloys were limited to tensile tests at room and elevated temperatures and to creep tests at temperatures of 300 to 600 degrees Fahr. (150 to 315 degrees Cent.).

THE question of grain-size is one which is receiving a large amount of attention at the present time. Very little information is available in the literature, however, concerning the effects of this factor on the various high temperature properties of metals. Yet, from theoretical considerations, it is believed that grain-size should exert an appreciable influence on the high temperature behavior of metals and, especially, insofar as the creep characteristics are concerned.

A paper presented as part of the Grain-Size Symposium of the Sixteenth Annual Convention of the Society held in New York City the week of October 1, 1934. The authors are members of the Society and are respectively, Director of Engineering Research, University of Michigan and Research Engineer, of the same department, Ann Arbor, Michigan. Manuscript received June 29, 1934.

In this connection two types of grain-size are to be considered. One is that grain-size which is revealed through the McQuaid-Ehn test and which is often designated as the inherent grain-size; while the second is the grain-size produced through heat treatment and which is often called the actual grain-size.

In this paper both types of grain-size will be considered. For the McQuaid-Ehn, or inherent, grain-size investigation, two alloy steels of the pearlitic type were used, one of which was of the molybdenum type and the second, a manganese-molybdenum steel.

Table I
Chemical Composition, Heat Treatment, Brinell Hardness and Grain-Size of Two Alloy Steels

Designation	Composition C	Mn	Si	Mo	Heat Treatment Deg. Fahr.	Brinell Hardness	Grain-Size
C-Mo (F)	0.16	0.49	0.30	0.49	Ann. 1550	131	8
C-Mo (C)	0.16	0.47	0.23	0.42	Ann. 1550	126	4–5
MM9 (F)	0.15	1.28	0.29	0.28	N. 1725, D. 1200	152	7–8
MM9 (C)	0.15	1.25	0.19	0.25	N. 1725, D. 1200	140	4–5

For the investigations to determine the influence of actual grain-size or grain-size as produced through heat treatment, two nonferrous alloys were used. The nonferrous alloys were chosen for two reasons. First, a wide range of grain-sizes may be readily obtained, and second, variations in grain-size are not accompanied by other structural changes. When the grain-size of ferrous alloys is changed by means of heat treatment, certain other structural changes may also occur, especially in regard to the type or distribution of the pearlite.

McQuaid-Ehn (Inherent) Grain-Size

Two alloy steels were considered in this investigation, both of which contained approximately the same carbon content and differed in that one contained 0.50 per cent molybdenum and the other 1.25 per cent manganese-0.25 per cent molybdenum. The chemical composition, heat treatment, Brinell hardness and grain-size of these steels are given in Table I. All four steels were made in an electric furnace and were obtained from The Timken Steel and Tube Company. The necessary specimens were machined from 1-inch round rods and the heat treatment was done prior to machining.

The steels were subjected to short-time tensile and Charpy impact tests at room and elevated temperatures and to creep tests at 800, 1000 and 1200 degrees Fahr. (425, 540, 650 degrees Cent.).

Tensile Tests

Short-time tensile tests were conducted on these alloys at 80, 750, 900, 1000, 1100, 1200, 1300 and 1400 degrees Fahr. (25, 400, 485, 540, 595, 650, 705, 760 degrees Cent.). The values considered are tensile strength, yield stress, proportional limit, per cent elongation in 2 inches and per cent reduction of area. The yield stress is defined as that stress which produces a permanent deformation of 0.2 per cent of the gage length and is obtained from the stress-strain curves. The results obtained are given in Figs. 1 to 4, inclusive. Each reported value is the average of two tests.

The stress-strain curves from which the proportional limit and yield stress values were obtained were determined with an optical extensometer system which is sensitive to 2.8 millionths of an inch per inch of 2-inch gage section.

Strength Characteristics—Under this heading are considered the tensile strength, yield stress and proportional limit values. The results obtained from the carbon-molybdenum steel are given in Fig. 1 and those from the manganese-molybdenum alloy in Fig. 3.

With both of these steels it is observed that the coarse-grained materials possess the maximum tensile strength values over the major portion of the temperature range considered. The difference between the two appears to be a maximum at temperatures of 750 to 1000 degrees Fahr. (400-540 degrees Cent.). At the higher temperatures, that is 1200, 1300 and 1400 degrees Fahr. (650, 705, 760 degrees Cent.), this value is approximately the same for both grain-sizes considered.

In the case of the yield stress values of the manganese-molybdenum steel, the coarse-grained material is again superior to the fine-grained over the major portion of the temperature range with the difference between the two being fairly constant up to 1200 degrees Fahr. (650 degrees Cent.) and being very slight at 1300 and 1400 degrees Fahr. (705, 760 degrees Cent.). With the carbon-molybdenum steel, however, a uniform difference does not exist between the fine- and coarse-grained materials. At temperatures from 80 to 900 degrees Fahr. (25-480 degrees Cent.) the fine-grained material is superior, while at temperatures of 1000 degrees Fahr. (540 degrees Cent.) and greater, the coarse-grained steel possesses the higher value.

Neither does grain-size produce a uniform difference in the

proportional limit values of these two steels. With the manganese-molybdenum alloy, the coarse-grained material is either equal to or greater than the fine-grained, but the difference between the two is in all cases slight. With the carbon-molybdenum steel, the fine-

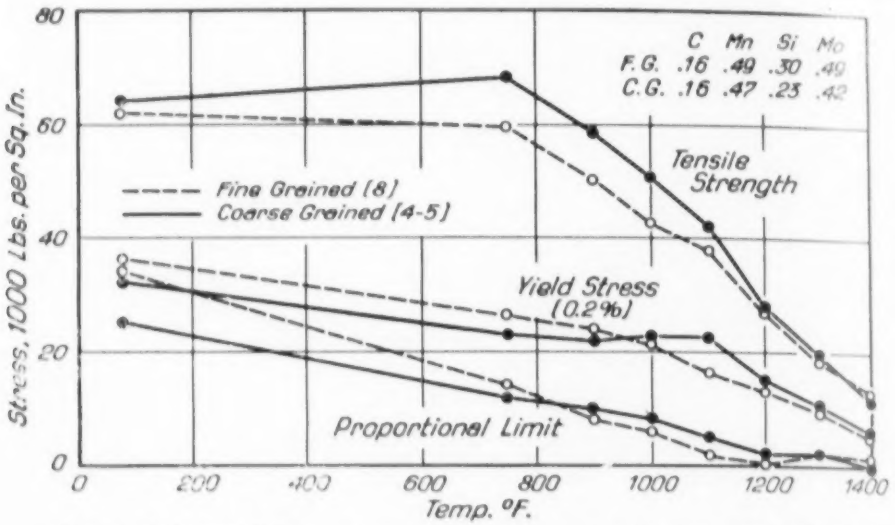


Fig. 1—Influence of Temperature and Grain-Size on the Strength Characteristics of Steel C-Mo.

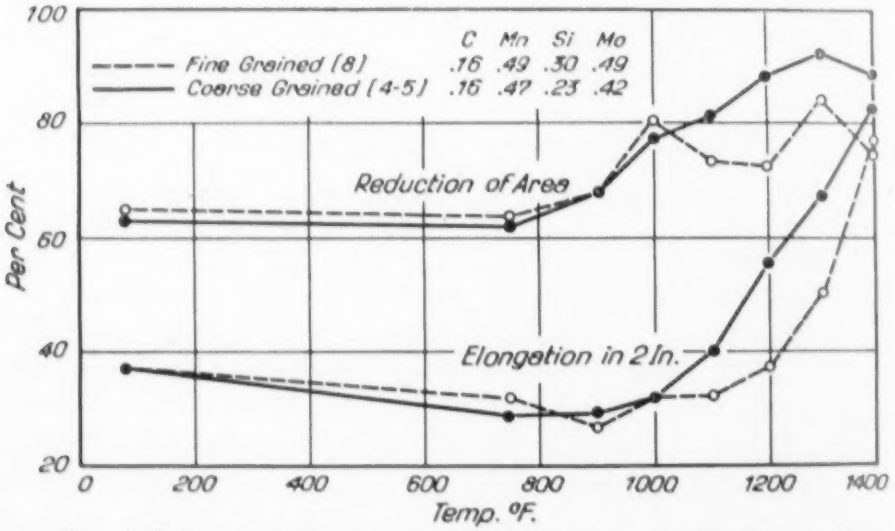


Fig. 2—Influence of Temperature and Grain-Size on the Ductility of Steel C-Mo.

grained material possesses the higher values at the lower temperatures, with the difference between the two being especially marked at room temperature. At temperatures of 900 degrees Fahr. (480 degrees Cent.) and higher, the coarse-grained material is in most

cases superior to the fine-grained. The difference between the two, however, is slight, especially at 1200, 1300 and 1400 degrees Fahr. (650, 705, 760 degrees Cent.).

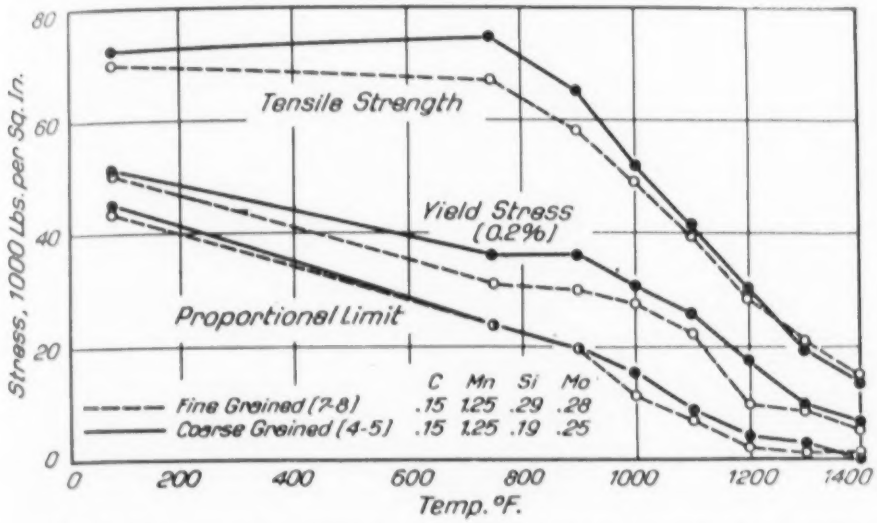

Stress, 1000 Lbs. per Sq. In.
Tensile Strength
Yield Stress (0.2%)
Proportional Limit
C Mn Si Mo
Fine Grained (7-8) .15 1.25 .29 .28
Coarse Grained (4-5) .15 1.25 .19 .25
Temp. °F.

Fig. 3—Influence of Temperature and Grain-Size on the Strength Characteristics of Steel MM9.

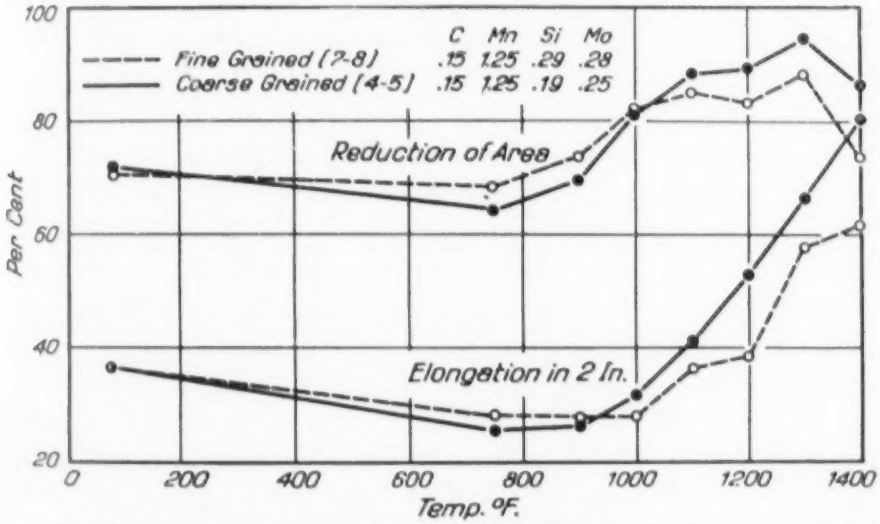


Fig. 4—Influence of Temperature and Grain-Size on the Ductility of Steel MM9.

Ductility Characteristics—The values considered in this connection are per cent elongation in 2 inches and per cent reduction of area. The values obtained from the carbon-molybdenum steel are given in Fig. 2 and those from the manganese-molybdenum alloy in Fig. 4.

In respect to the per cent elongation, the fine- and coarse-grained materials give approximately the same value at room temperature. Also, in both cases, the fine-grained material is superior at the lower temperatures and the coarse-grained at the higher ones. In the case of the carbon-molybdenum steel, the dividing temperature is at 1000 degrees Fahr. (540 degrees Cent.), while with the manganese-molybdenum alloy it lies between 900 and 1000 degrees Fahr. (485, 540 degrees Cent.).

Grain-size does not have a uniform influence on the reduction of area values. At room temperature the fine-grained material produces the maximum value in the case of the carbon-molybdenum alloy, while the coarse-grained material has a slightly higher value in the case of the manganese-molybdenum steel. Both steels are similar, however, in that at temperatures up to approximately 1000 degrees Fahr. (540 degrees Cent.) the fine-grained material possesses the higher value, while at temperatures above 1000 degrees Fahr. (540 degrees Cent.) the coarse-grained steels have the higher values.

Hot Impact Tests

Charpy impact tests were conducted on all of these materials at temperatures of 80, 500, 600, 750, 900, 1000, 1100 and 1200 degrees Fahr. (25, 260, 315, 400, 485, 540, 595, 650 degrees Cent.). A key-hole notch was used and all tests were conducted in triplicate. In one series of tests the specimens were held at temperature one hour before testing, while in the second series, the specimens were held at temperature 1000 hours before testing. In all cases, the specimens were heated in an external furnace and rapidly transferred to the machine and broken. The time required for this operation was approximately three seconds. The results obtained are given in Figs. 5 to 8, inclusive.

First, considering the one hour test, it is observed that at room temperature the fine-grained material possesses the higher resistance to impact. In the case of the carbon-molybdenum steel, the fine-grained material is superior at all temperatures except at 1000 degrees Fahr. (540 degrees Cent.). At this temperature grain-size does not appreciably influence the impact values. With the manganese-molybdenum steel the fine-grained material possesses values equal to or greater than those obtained from the coarse-grained steels at temperatures up to 800 degrees Fahr. (425 degrees Cent.). At the higher temperatures, however, the coarse-grained

steel is the superior one. In the majority of cases, however, the difference between the values is not great.

The results obtained from the specimens which were held 1000 hours at temperature before testing show more uniform differences, especially in the case of the manganese-molybdenum alloy. With both

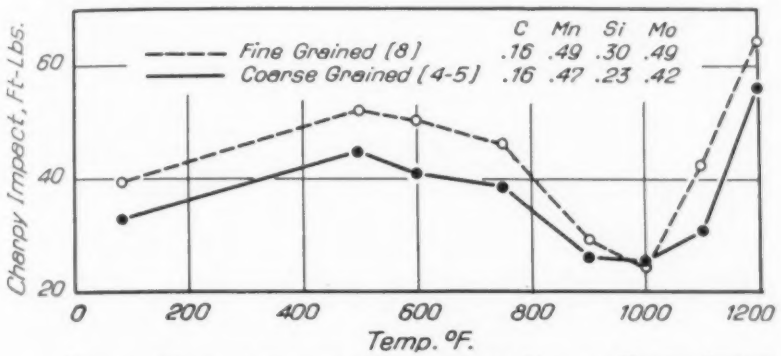


Fig. 5—Effect of Temperature and Grain-Size on Charpy Impact Resistance of Steel C-Mo. Specimens Held 1 Hour at Temperature Before Testing.

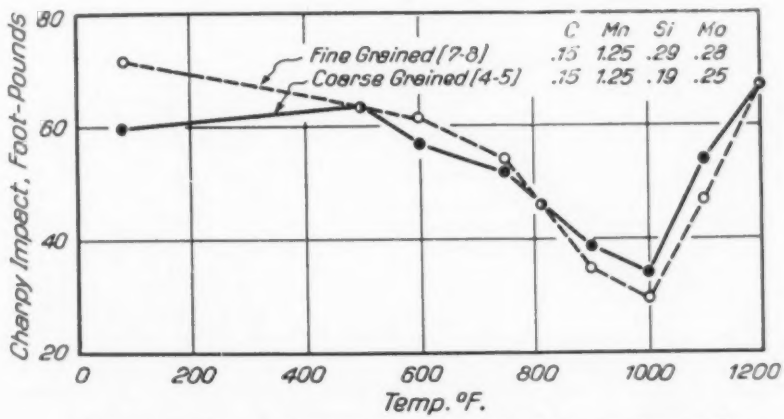


Fig. 6—Effect of Temperature and Grain-Size on Charpy Impact Resistance of Steel MM9. Specimens Held 1 Hour at Temperature Before Testing.

steels the fine-grained material possesses the maximum impact resistance over the entire temperature range considered. While tests of this type were also conducted at 1200 degrees Fahr. (650 degrees Cent.) the results obtained are not included as they were appreciably influenced by oxidation and structural changes which occurred during this long-time heating period.

Creep Characteristics

The creep characteristics of these materials were determined

at 800, 1000 and 1200 degrees Fahr. (425, 540, 650 degrees Cent.). The so-called single-step method of loading was used with four or more stresses being employed for each steel at each temperature and each test being continued for at least 500 hours. An optical extensometer was used for determining the elongation which is sensitive

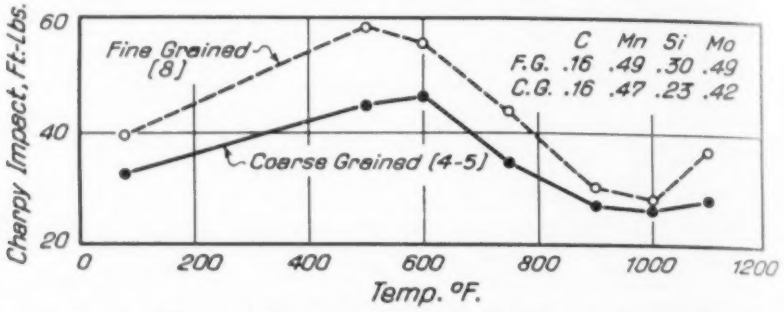


Fig. 7—Effect of Temperature and Grain-Size on Charpy Impact Resistance of Steel C-Mo. Specimens Held 1000 Hours at Temperature Before Testing.

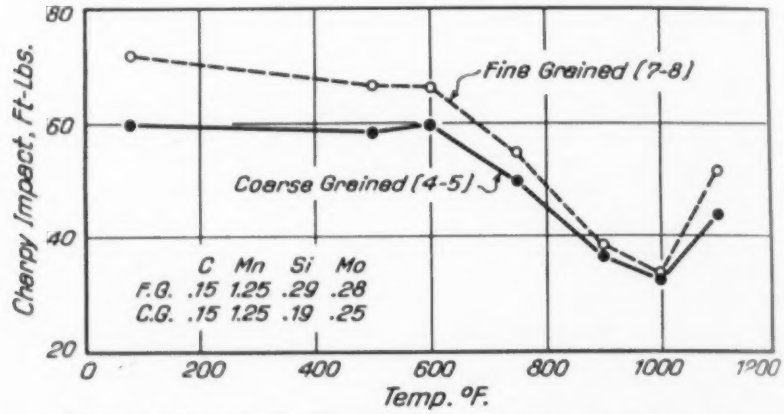


Fig. 8—Effect of Temperature and Grain-Size on Charpy Impact Resistance of Steel MM9. Specimens Held 1000 Hours at Temperature Before Testing.

to 2.8 millionths of an inch per inch of 2 inch gage section. The apparatus employed for these tests has been previously described in the literature.[1]

The results obtained are given in Table II and in Figs. 9 and 10. In the figures the rates of creep and the corresponding stresses are plotted to logarithmic coordinates. This method of plotting is useful in that under many conditions a straight-line relationship results. The rate of creep corresponding to any stress, within certain limits, or the stress corresponding to any rate of creep can thus be

[1]A. E. White, C. L. Clark and L. Thomassen, "An Apparatus for the Determination of Creep at Elevated Temperatures," Fuels and Steam Power, *Transactions*, American Society for Mechanical Engineers, Vol. 52, 1931, No. 27, p. 347.

estimated. The values given in the table were obtained from these figures.

While the results given in the table are expressed in terms of the stresses required for creep rates as low as 1 per cent per 100,000 hours, no tests have been conducted for time periods even approaching this. It is customary to assume that a creep rate of 0.01 per cent per 1000 hours is equivalent to a creep rate of 1 per cent per 100,000 hours.

Table II

Influence of Grain-Size on the Creep Characteristics of Steels C-Mo and MM9 at 800, 1000 and 1200 Degrees Fahr.

Designation	Temperature Deg. Fahr.	Stress for Creep Rate of 1 Per Cent in Designated Time Period		
		100,000 Hrs.	10,000 Hrs.	1,000 Hrs.
C-Mo (F)	800	21,500	27,000	34,000
C-Mo (C)	800	15,500	26,000	44,500
C-Mo (F)	1000	9,600	15,800	26,000
C-Mo (C)	1000	10,700	17,800	29,200
C-Mo (F)	1200	260	1,400	3,400
C-Mo (C)	1200	480	2,000	4,050
MM9 (F)	800	25,500	34,000	45,000
MM9 (C)	800	27,000	38,500	55,000
MM9 (F)	1000	4,650	9,800	21,000
MM9 (C)	1000	7,400	14,000	18,250
MM9 (F)	1200	405	970	2,325
MM9 (C)	1200	460	1,075	2,500

F = Fine-grained (7-8)
C = Coarse-grained (4-5)

On the basis of theoretical considerations, it is believed that the influence of grain-size on the creep characteristics should vary, depending upon the temperature being considered. It is the writers' belief that at temperatures below the so-called equi-cohesive temperature, or lowest temperature of recrystallization, a fine-grained material should possess the maximum creep resistance, while at temperatures above, the coarse-grained material should be superior. Since time is also a deciding factor, this condition should be especially true when low creep rates are being considered.

Referring to the table and figures, it will be observed that this condition does exist. For example, with the carbon-molybdenum steel, the fine-grained material is superior to the coarse at 800 degrees Fahr. (425 degrees Cent.), with the difference being the most marked in the case of the stress required for a creep rate of 1 per cent per 100,000 hours. In the case of the largest creep rate considered, 1 per cent per 1000 hours, the coarse-grained material was superior, but, because of this relatively rapid rate of creep, it is believed that

the tensile, rather than the creep characteristics, are the deciding factor under this condition. At the higher temperatures, 1000 and 1200 degrees Fahr. (540, 650 degrees Cent.), the coarse-grained material possesses a higher creep resistance than the fine-grained. In other words, for this steel it is believed that the equi-cohesive temperature lies somewhere between 800 and 1000 degrees Fahr. (425, 540 degrees Cent.).

The same general conditions as noted above exist in the case of the manganese-molybdenum steel. With this alloy, however, the

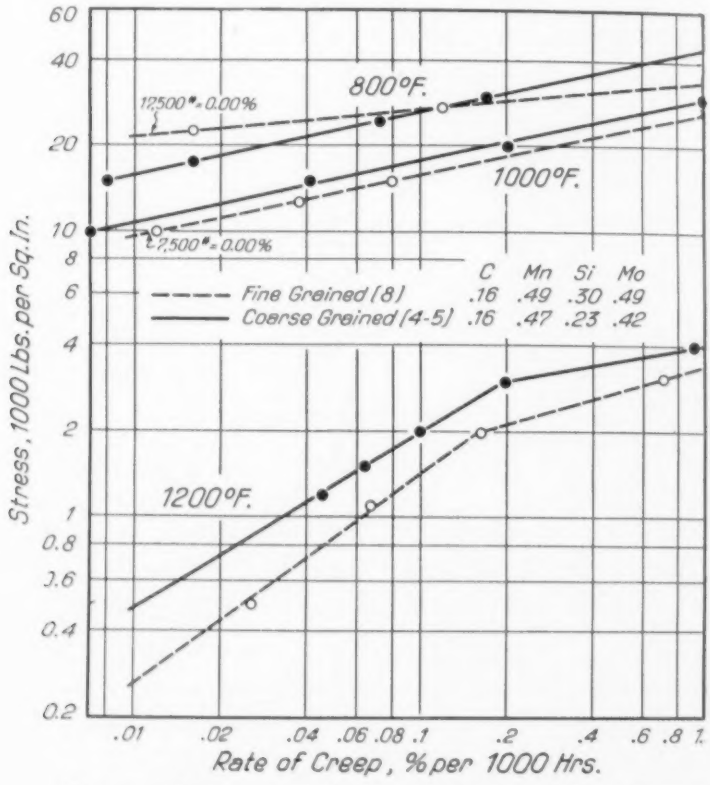


Fig. 9—Influence of Grain Size on the Creep Stress Curves for Steel C-Mo at 800, 1000 and 1200 Degrees Fahr.

coarse-grained material possesses the higher creep resistance at each of the three temperatures considered. This would indicate either that the equi-cohesive temperature of this alloy lies below 800 degrees Fahr. (425 degrees Cent.), or that other factors, such as composition or a difference in the strain-hardening characteristics of the two grain-sizes, are appreciably influencing the results.

In Figs. 9 and 10 it is apparent that the log-log lines at any given temperature obtained from the steels of different grain-sizes

are generally not parallel. With both the molybdenum and the manganese-molybdenum steels at 800 degrees Fahr. (425 degrees Cent.), the coarse-grained material gives the greater slope, while with the manganese-molybdenum alloy the fine-grained material has the greater slope at 1000 degrees Fahr. (540 degrees Cent.). It is the writers' belief that these variations in the slopes are due to

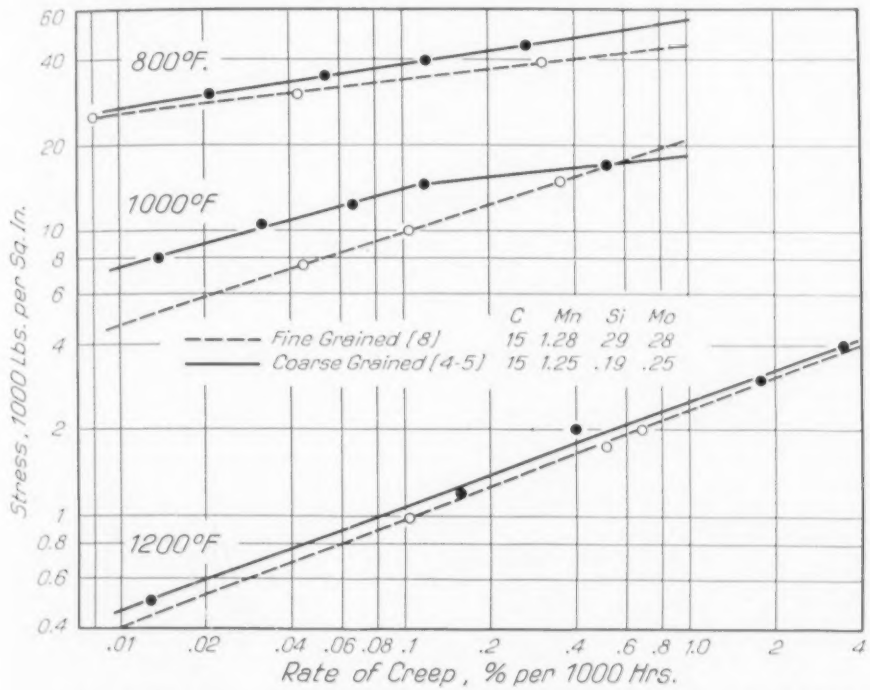


Fig. 10—Influence of Grain Size on the Creep-Stress Curves for Steel MM9 at 800, 1000 and 1200 Degrees Fahr.

differences in the strain-hardening characteristics of the materials, with the greater slope indicating a greater susceptibility to strain-hardening. In fact, it is believed that many of the physical effects produced by varying grain-size may be largely caused by the differences in the strain-hardening characteristics of fine- and coarse-grained steels.

High Temperature Stability

The structural stability of metals at elevated temperatures can be determined by subjecting the completed creep specimen to tensile and impact tests and a metallographic examination. As four specimens are used for each creep test at each temperature, two of the completed specimens were subjected to tensile tests and the remaining two to the impact tests and the metallographic examination.

Tensile Tests—The results obtained from the tensile tests on the completed creep specimens at 800, 1000 and 1200 degrees Fahr. (425, 540, 650 degrees Cent.) are given in Table III. Results obtained from the original material are also included for comparative purposes.

The creep tests at 800 and 1000 degrees Fahr. (425, 540 degrees Cent.) have had no appreciable influence on the tensile characteristics of any of these four steels. In certain cases the creep tests at 800 degrees Fahr. (425 degrees Cent.) have produced a slight increase in the resulting strength characteristics and this is the more marked in the yield stress values obtained from the coarse-grained steels. This is believed to be due to the strain-hardening which occurred during the creep tests at this temperature.

The creep tests at 1200 degrees Fahr. (650 degrees Cent.), however, have had a more marked influence. In practically every case, the strength characteristics have been somewhat decreased with the tensile strength values, as a rule, showing the more marked changes. In no cases, however, are the differences exceedingly

Table III

Influence of Grain-Size on the Tensile Properties of C-Mo and MM9 Steels Before and After Creep Tests at 800, 1000 and 1200 Degrees Fahr.

Specimen Number		Creep Tests: Temperature Deg. Fahr.	Creep Tests: Stress Pounds Per Square Inch	Tensile Strength Pounds Per Square Inch	Yield Stress Pounds Per Square Inch	Proportional Limit Pounds Per Square Inch	Elongation Per Cent in 2 Inches	Reduction of Area Per Cent
CMo-0	(F)	85	.....	62,475	36,000	34,000	37.25	64.7
CMo8-15	(F)	800	15,000	60,000	34,000	30,000	32.5	64.2
CMo8-27	(F)	800	27,500	61,250	40,000	32,000	33.5	64.9
CMo10-7	(F)	1000	7,500	60,650	38,000	30,000	37.0	66.1
CMo10-15	(F)	1000	15,000	63,000	40,000	32,000	31.0	65.4
CMo12-15	(F)	1200	1,500	59,550	35,000	30,000	33.0	71.5
CMo-0	(C)	85	.....	64,050	32,500	25,000	37.0	62.7
CMo8-17	(C)	800	17,500	62,200	40,000	32,500	33.0	62.6
CMo8-30	(C)	800	30,000	63,550	42,000	30,000	32.0	62.4
CMo10-15	(C)	1000	15,000	61,800	35,000	27,500	33.5	64.5
CMo10-30	(C)	1000	30,000	65,200	43,750	35,000	26.0	62.8
CMo12-15	(C)	1200	1,500	52,750	34,000	27,500	34.0	79.7
CMo12-4	(C)	1200	4,000	54,000	31,000	22,500	33.0	73.7
MM9-0	(F)	85	.....	69,825	50,000	43,750	36.5	70.4
MM9(8)-25	(F)	800	25,000	70,800	47,500	37,500	36.0	69.9
MM9(8)-39	(F)	800	39,000	71,300	52,500	40,000	32.5	70.0
MM9(10)-5	(F)	1000	5,000	70,500	50,000	45,000	34.0	71.5
MM9(10)-10	(F)	1000	10,000	70,400	47,500	42,500	35.5	69.6
MM9(12)-2	(F)	1200	1,750	60,400	45,000	30,000	31.0	72.8
MM9-0	(C)	85	.....	72,125	51,250	45,000	36.25	71.85
MM9(8)-35	(C)	800	35,000	76,450	57,500	45,000	32.0	71.3
MM9(8)-45	(C)	800	45,000	76,100	57,500	42,500	32.0	72.4
MM9(10)-10	(C)	1000	10,500	71,250	50,000	46,000	33.0	70.8
MM9(10)-17	(C)	1000	17,000	70,300	48,000	45,000	28.5	71.4
MM9(12)-2	(C)	1200	2,000	59,550	40,000	30,000	33.0	76.3
MM9(12)-4	(C)	1200	4,080	58,900	37,500	27,500	34.0	77.6

marked. With both of the compositions considered, the change in the coarse-grained steel was the greater.

Impact Tests—For these tests Izod specimens were used in order that two specimens might be obtained from each completed creep specimen. Since 0.505 inch diameter specimens were used for the creep tests, the resulting impact specimens were smaller than standard.

Table IV

Influence of Grain-Size on the Impact Resistance of Steels C-Mo and MM9 Before and After Creep Tests at 800, 1000 and 1200 Degrees Fahr.

Specimen Number		Temperature Degrees Fahr.	Stress Lb./Sq. In.	Duration Hours	Izod Impact* Foot-Pounds
CMo-0	(F)	85	.....	...	30.5
CMo8-10	(F)	800	10,000	500	29.75
CMo8-22	(F)	800	22,500	575	18.50
CMo10-10	(F)	1000	10,000	507	27.50
CMo10-12	(F)	1000	12,500	500	25.25
CMo12-1	(F)	1200	1,000	480	45.00
CMo12-3	(F)	1200	3,100	500	42.50
CMo-0	(C)	85	.....	...	47.0
CMo8-15	(C)	800	15,000	570	41.0
CMo8-24	(C)	800	24,250	570	33.0
CMo10-10	(C)	1000	10,000	550	52.0
CMo10-20	(C)	1000	20,000	500	38.5
CMo12-12	(C)	1200	1,200	620	82.0
CMo12-30	(C)	1200	3,000	500	66.0
MM9-0	(F)	85	.....	...	93.0
MM9(8)-20	(F)	800	20,000	515	92.5
MM9(8)-30	(F)	800	30,000	500	93.0
MM9(10)-7	(F)	1000	7,500	590	90.0
MM9(10)-15	(F)	1000	15,000	500	91.5
MM9(12)-1	(F)	1200	1,100	530	94.0
MM9(12)-4	(F)	1200	4,000	500	94.0
MM9-0	(C)	85	.....	...	96.5
MM9(8)-40	(C)	800	40,000	500	95.0
MM9(10)-8	(C)	1000	8,000	570	94.0
MM9(10)-14	(C)	1000	14,750	600	91.5
MM9(12)-1	(C)	1200	1,100	515	89.5
MM9(12)-3	(C)	1200	3,000	500	91.0

*Average of two tests.
Note: Specimens used were smaller than standard.

The results obtained from the impact tests are given in Table IV. Values obtained from the original material are also included for comparative purposes. All reported values are the average of two tests.

The results indicate that in no case have the creep tests at 800 or 1000 degrees Fahr. (425-540 degrees Cent.) had an appreciable influence on the resulting impact resistance of these steels. The same is likewise true with the creep tests at 1200 degrees Fahr. (650 degrees Cent.) in the case of the manganese-molybdenum alloy. With the molybdenum steel, however, the creep tests at 1200 degrees Fahr. (650 degrees Cent.) have caused a considerable increase in

the resulting impact resistance. With the fine-grained material this increase is of the order of 50 per cent, while with the coarse-grained material the resistance is nearly doubled.

Metallographic Examination—The metallographic structures of each steel were examined, both before and after the creep tests. The results indicate that the creep tests at 800 and 1000 degrees Fahr. (425-540 degrees Cent.) had no apparent influence on the structure. Those at 1200 degrees Fahr. (650 degrees Cent.) did produce a structural change in that the pearlite was more highly spheroidized. Even in the original condition, however, none of these steels possessed a distinctly laminar pearlite and it could best be defined by the term "sorbitic pearlite."

Conclusions

On the basis of the above series of tests it must be concluded that the McQuaid-Ehn (inherent) grain-size does appreciably influence the high temperature characteristics of steel. On the basis of the tensile test results the coarse-grained materials in general possess the superior strength characteristics. There are exceptions to this statement in the case of certain of the yield stress and proportional limit values of the molybdenum steel. At temperatures from 80 to 900 or 1000 degrees Fahr. (25, 480, 540 degrees Cent.) the fine-grained materials possess the greater ductility, while at temperatures above 1000 degrees Fahr. (540 degrees Cent.) the coarse-grained steels are superior in this respect.

The fine-grained steels also possess the maximum impact resistance over the major portion of the temperature range from 80 to 1200 degrees Fahr. (25 to 650 degrees Cent.). This is true regardless of whether the specimens are held at temperature one hour or 1000 hours before testing. In the long heating tests there are no exceptions to this statement, while in the one hour tests there is, in that the coarse-grained manganese-molybdenum steel possesses the maximum values at temperatures from 900 to 1200 degrees Fahr. (480-650 degrees Cent.).

Insofar as the creep characteristics are concerned the fine-grained steels are superior at the lower temperatures and the coarse-grained steels at the higher ones. It is the writers' belief that, in general, the dividing temperature coincides with the lowest temperature of recrystallization or the equi-cohesive temperature. On this basis, the equi-cohesive temperature of the molybdenum steel would

lie between 800 and 1000 degrees Fahr. (425-540 degrees Cent.), while that of the manganese-molybdenum alloy would be below 800 degrees Fahr. (425 degrees Cent.). Other factors, such as composition and marked differences in strain-hardening characteristics, may appreciably influence this relationship.

Grain-Size Produced by Heat Treatment (Actual)

For this investigation two nonferrous alloys were used, one of which was of the 77 copper, 22 zinc, 1 tin and the other the 59 copper, 40 zinc, 1 tin type. As stated previously, nonferrous alloys were selected because variations in grain-size could be readily obtained and these variations are not as likely to introduce other structural variables as in ferrous alloys. It is believed, however, that the findings obtained will apply to ferrous as well as nonferrous materials. The designation, composition and grain-size of these alloys are given in Table V.

Table V
Chemical Composition of Nonferrous Alloys With Varying Grain-Size

Designation	Alloy	Chemical Composition Per Cent Copper	Zinc	Tin	Grain-Size Millimeters
E-2	77-22-1	77.26	21.61	1.18	0.020
E-20	77-22-1	77.24	21.77	1.02	0.045
E-3	59-40-1	58.79	40.43	0.88	0.025
E-24	59-40-1	60.08	38.72	0.83	0.045

These alloys were received from commercial sources and were furnished in 0.75-inch rods. The materials were subjected to short-time tensile tests at both room and elevated temperatures and to creep tests at temperatures ranging from 300 to 600 degrees Fahr. (150-315 degrees Cent.).

Tensile Properties

Tensile tests were conducted at 80, 400, 600 and 800 degrees Fahr. (25, 205, 315, 425 degrees Cent.) on each of these four alloys. The values considered were tensile strength, proportional limit, per cent elongation in 2 inches and per cent reduction of area. The results obtained are shown in Figs. 11 and 12.

First considering the strength characteristics, that is, the tensile strength and proportional limit values, it is observed that the

fine-grained material possesses the maximum values in all cases with the difference between the two tending to be greater at 400 degrees Fahr. (205 degrees Cent.).

The influence of grain-size on the ductility, Fig. 12, is not as uniform. With the 59-40-1 alloy the coarse-grained material gives

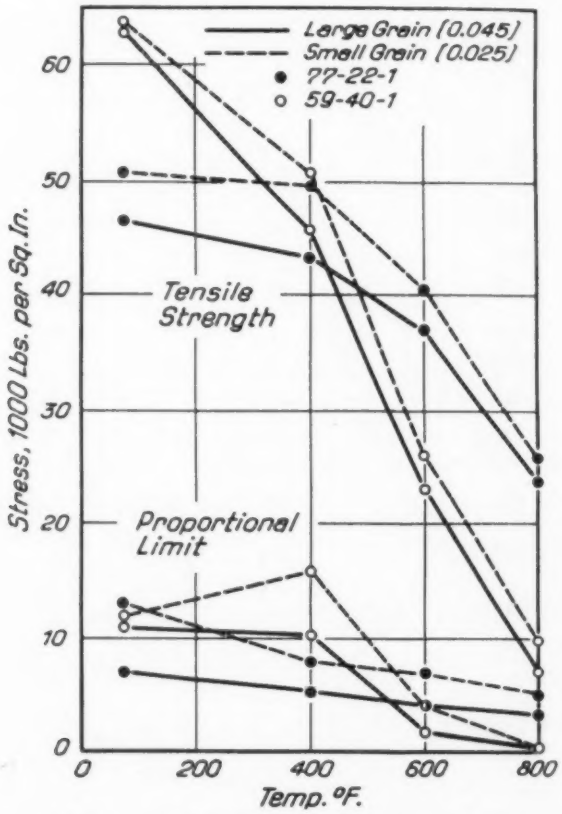


Fig. 11—Influence of Temperature and Actual Grain-Size on the Strength of Two Non-Ferrous Alloys.

the higher elongation value at every temperature except 400 degrees Fahr. (205 degrees Cent.), at which temperature grain-size does not influence the resulting value. With the 77-21-1 alloy the maximum values are possessed by the coarse-grained structure at all temperatures except 800 degrees Fahr. (425 degrees Cent.). In the case of the reduction of area values, the fine-grained 77-22-1 alloy possesses the higher value over the entire temperature range, while with the 59-40-1 alloy the same is true only at 400 degrees Fahr. (205 degrees Cent.). At the higher temperatures the coarse-grained structure is superior while at room temperature the two give the same value.

Creep Characteristics

The creep characteristics of the 77-22-1 alloy were determined at 400 and 600 degrees Fahr. (205, 315 degrees Cent.) and those of the 59-40-1 alloy at 300 and 400 degrees Fahr. (150, 205 degrees Cent.). These particular temperatures were selected as it was desired to conduct tests, both below and above the lowest temperature of recrystallization or equi-cohesive temperature. Previous work has indicated that this temperature would lie between 400 and 600 degrees Fahr. (205, 315 degrees Cent.) for the 77-22-1 alloy and between 300 and 400 degrees Fahr. (150, 205 degrees Cent.) for the 59-40-1 material. The results obtained are given in Fig. 13.

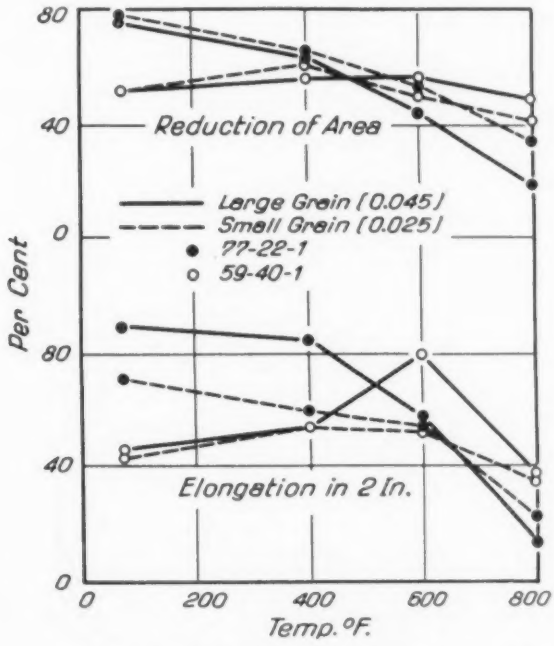


Fig. 12—Influence of Temperature and Actual Grain-Size on the Ductility of Two Non-Ferrous Alloys.

The writers have previously stated on several occasions that the influence of actual grain-size on the creep characteristics would vary, depending upon the testing temperature. Their reasons for this statement are based on the following facts. At room temperature fracture resulting from tension and thus the greater share of deformation occurs through the grain. In other words, at these temperatures the grain boundaries are stronger and more able to withstand deformation than the grains themselves and, therefore,

a fine-grained material will possess the greater strength because of the greater relative amount of grain boundaries present.

As the temperature is increased, however, the grain boundaries weaken at a more rapid rate than the crystals and a temperature is reached at which fracture proceeds around rather than through the

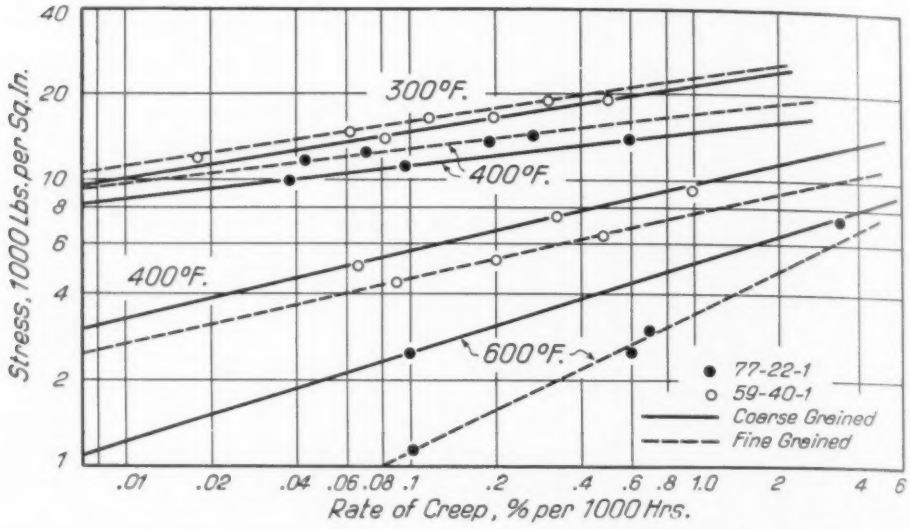


Fig. 13—Creep Stress Diagram for Copper-Zinc-Tin Alloys Showing the Influence of Grain-Size on Rate of Creep.

grains. At this temperature a coarse-grained material will possess the greatest strength because of the presence of a larger amount of the stronger phase, the crystalline. There is still considerable question as to the location of this dividing temperature. The writers believe that it is the lowest recrystallization temperature or the equicohesive temperature.

The data given in Fig. 13 support this statment. At 400 degrees Fahr. (205 degrees Cent.) the fine-grained 77-22-1 alloy is more resistant to continuous creep than the coarse-grained material, while at 600 degrees Fahr. (315 degrees Cent.) the coarse-grained alloy offers the greater resistance. Since the lowest temperature of recrystallization of this material has been found to be between 400 and 600 degrees Fahr. (205-315 degrees Cent.) the hypothesis is supported. Likewise, with the 59-40-1 alloy the fine-grained material is the more resistant at 300 degrees Fahr. 150 degrees Cent.) and the coarse-grained material is superior at 400 degrees Fahr. (205 degrees Cent.). Again this change is closely related to the lowest temperature of recrystallization.

Conclusions

On the basis of the above tests, it is to be concluded that actual grain-size, that is, grain-size produced through heat treatment, does have an influence on the resulting high temperature characteristics. Insofar as the tensile properties are concerned, it is observed that the fine-grained material possesses the greater strength properties over the entire temperature range considered. The influence of grain-size on the ductility is not as uniform, but, in general, the coarse-grained material tends to give the greater elongation values, especially at the lower temperatures, while the fine-grained material gives, in the majority of cases, the greater reduction of area values.

The results from the creep tests indicate that at temperatures below the lowest temperature of recrystallization fine-grained materials possess the greater creep resistance, while at temperatures above the coarse-grained material has the greater resistance to creep.

Inherent Versus Actual Grain-Size

Results which have been considered should give an indication at least of the relative influence of grain-size as determined by the McQuaid-Ehn test (inherent) and that produced through heat treatment (actual) on the resulting high temperature characteristics.

First considering the creep characteristics, it is observed that regardless of which type of grain-size is being considered (actual or inherent), the maximum creep resistance at the lower temperatures is obtained with fine-grained materials, while at the higher temperatures a coarse-grained alloy is required for the greater creep resistance. In the case of the nonferrous alloys (actual grain-size), this dividing temperature has been related to the lowest temperature of recrystallization or equi-cohesive temperature. While no information is available concerning the lowest temperature of recrystallization of the two ferrous alloys (inherent grain-size), it is believed that this would again be the dividing temperature.

Both types of grain-size do not exert the same influence on the tensile properties at either room or elevated temperatures. In the case of the nonferrous alloys (actual grain-size), it was found that the maximum strength and minimum elongation values were, in general, possessed by the fine-grained structures. On the basis of inherent grain-size, however, the maximum strength characteristics were, in general, possessed by the coarse-grained steels. There were

certain exceptions to this, especially in connection with the yield stress and proportional limit values of the molybdenum steel at temperatures below 800 to 900 degrees Fahr. (425, 485 degrees Cent.). The influence of inherent grain-size on ductility was also different than that of actual grain-size. With both of the ferrous alloys considered, the fine-grained steels gave the maximum values at temperatures up to 800 to 1000 degrees Fahr. (425 to 540 degrees Cent.), while the coarse-grained steels were superior at the more elevated temperatures.

It appears, therefore, that on the basis of these tests, at least, both types of grain-size exert the same influence on the creep characteristics, but that their effect is different on the tensile properties at either room or elevated temperatures.

In this connection, it should be noted that at least insofar as steels are concerned, the inherent and actual grain-sizes are to some extent related. The actual grain-size that is manifested in the structure usually depends upon the inherent grain-size. For instance, if a coarse-grained and fine-grained steel are given an identical normalizing, annealing or quenching treatment, the difference in the actual structural grain-size of the steel is traced to the difference in inherent grain-size. In other words, the inherent grain-size does not depend upon the actual grain-size, but the actual grain-size may be a consequence of the inherent grain-size.

While the coarse-grained steels possess the superior creep characteristics at the more elevated temperatures, the difference between the fine- and coarse-grained steels often is not sufficiently large to be of great commercial importance. In such cases greater emphasis should perhaps be placed on the other mechanical properties which grain-size is known to influence.

Acknowledgments

The authors desire to express their appreciation to The Timken Steel and Tube Company and The Detroit Edison Company for their financial assistance which has made these investigations possible and for their permission to publish the findings. Particularly do they wish to express their gratitude to F. J. Griffiths and R. L. Wilson of The Timken Steel and Tube Company and to C. F. Hirshfeld, J. W. Parker and J. H. Walker of The Detroit Edison Company.

DISCUSSION

Writen Discussion: By H. D. Newell, chief metallurgist, the Babcock & Wilcox Tube Co., Beaver Falls, Pa.

This paper is further evidence of the continued interest in the characteristics of metals intended for elevated temperatures, and is another intresting contribution to the subject in which the authors have long been actively identified.

The writer's interpretation of the subject matter is that, in the case of pearlitic steels, the slight advantage in strength characteristics shown for the

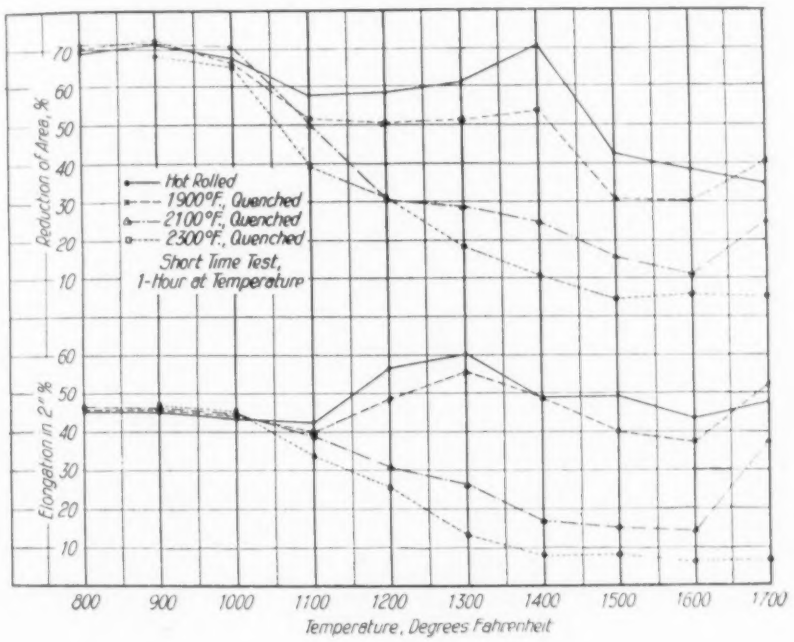


Fig. 1—Effect of Heat Treatment on the Ductility of 0.05 Per Cent Carbon. "18-8" Alloy at Elevated Temperatures. Short Time Test. 1 Hour at Temperature.

coarse-grained steels is partly offset by the slightly lower impact values and ductility of the large-grained steels, up to about 1000 degrees Fahr. On the basis of creep, the coarse-grained steels are preferable. In the case of the nonferrous alloys tested, the fine-grained alloys possess greater strength characteristics, but the coarse-grained ones exhibit greater elongation values.

In working with chromium-nickel austenitic steels which are structurally similar to the alpha brass alloy 77-22-1, the writer found exactly the reverse to be true for elongation values when testing at higher temperatures than covered by the authors. In this case, fine-grained, low carbon 18-8 steel showed both greater short-time strength and much better ductility than coarse-grained alloy of exactly identical composition. For example, one inch round, heat treated to produce four grain sizes from fine to very coarse, gave values shown in Table I and Fig. 1 of this discussion.

Table 1

Effect of Heat Treatment on Physical Properties of 0.05 Per Cent Carbon, 18-8 Alloy, at Elevated Temperatures, Short Time Test, One Hour at Temperature

Temp. of Test Deg. Fahr.	Hot-Rolled	1900 Deg. Fahr. Quenched	2100 Deg. Fahr. Quenched	2300 Deg. Fahr. Quenched
		Ultimate Strength		
800	60,710	58,440	56,690	.....
900	58,310	55,345	54,252	51,848
1000	54,245	51,330	50,700	48,800
1100	47,902	46,303	44,705	41,600
1200	38,261	37,512	35,314	35,589
1300	30,819	28,821	29,040	25,750
1400	32,750	29,100	28,750	21,050
1500	18,530	17,882	15,580	15,934
1600	13,650	11,900	9,220	14,700
1700	9,950	9,323	8,542	7,592
		Reduction of Area in Per Cent		
800	69.3	70.6	71.4	....
900	71.1	72.3	71.4	68.2
1000	67.5	66.5	70.6	65.4
1100	57.5	51.7	50.6	39.4
1200	58.8	50.6	30.8	31.2
1300	61.1	51.1	28.9	13.0
1400	71.0	54.0	24.8	10.4
1500	42.8	31.2	15.9	5.0
1600	38.8	30.5	11.9	5.9
1700	35.0	40.8	25.8	5.8
		Per Cent Elongation in 2 Inches		
800	45.5	46.0	46.5	....
900	45.0	46.0	46.0	47.0
1000	43.5	44.0	44.5	45.0
1100	42.5	40.0	39.0	34.0
1200	57.0	49.0	31.0	26.0
1300	60.5	56.0	27.0	14.0
1400	49.0	49.0	17.5	8.5
1500	49.5	40.0	16.0	8.5
1600	43.5	37.5	14.5	6.5
1700	48.0	52.5	38.0	7.0

Table 1 and Fig. 1 of this discussion are reproduced from the writer's paper—"Influence of Grain Size on the Properties and Corrosion Resistance of the 18-8 Iron-Chromium-Nickel Alloy for Elevated Temperature Service." —TRANSACTIONS, American Society for Steel Treating, Vol. 19, 1931-1932, p. 673.

As to creep of 18-8 steel, the recent work of the Joint Research Committee[1] indicates somewhat superior values for fine-grained rolled alloy with much greater total elongation before fracture (high carbon alloy) when compared with large-grained cast alloy of identical composition. It may then be considered preferable, in the case of 18-8 alloys, to specify fine-grained metal for high temperature uses so as to obtain greater reliability.

However, it seems evident that the structural characteristics, precipitation phenomena and other properties of each composition of steel or alloy must be known before the optimum grain-size can be chosen for elevated temperature

[1]H. C. Cross, "High Temperature Tensile, Creep and Fatigue of Cast and Wrought High and Low-Carbon 18 Cr. 8 Ni. Steel From Split Heats," *Transactions*, American Society of Mechanical Engineers, July, 1934, p. 533.

service as no general rule will likely apply to all materials through all temperature ranges. Messrs. White and Clark are to be congratulated for extending our knowledge in this direction.

Written Discussion: By Alvin J. Herzig, metallurgist, Climax Molybdenum Co., Detroit.

The authors of this paper have concerned themselves with the subject of creep in metals for several years. In my opinion their work has opened to us more than any other a logical interpretation of the factors governing this property of metals. I am especially in accord with their views as brought out in previous papers that the equi-cohesive temperature and grain-size are of considerable significance in the study of creep. In the paper now under discussion they have combined the factors of grain-size and equi-cohesive temperature in further consideration of the characteristics of metals at elevated temperatures. In their previous papers the authors' term "grain-size" always meant "actual" grain-size. Today we do not use the term "grain-size" without qualifying it as "inherent" or "actual".

It occurs to the writer that the statement made by the authors that "inherent and actual grain-sizes are related" should be discussed further in connection with their high temperature problems. We shall eventually realize, I believe, that the McQuaid-Ehn test is of value to us only insofar as it gives us some measure of the grain growth characteristics of the metal. These characteristics are present by virtue of the composition and manufacturing practice and cannot be readily nor accurately predicted on the basis of chemical tests now in use. The grain growth characteristics, in turn, give us some notion as to whether or not we will be dealing more particularly with intra- or intergranular forces and energy.

Whatever our conception may be as to the cause of the restriction of grain growth in the so-called "fine-grained steel", the intercrystalline energy which controls the rates of structural changes and determines the properties of the yet hypothetical intercrystalline phase is the factor from which we primarily derive our final characteristics, and not the fine grain. If this were not so inherently fine-grained steels, like fine-grained nonferrous alloys, would be used for hardness and strength; whereas, in reality, inherently fine-grained steels develop toughness characteristics in spite of their fine grain and not because of it.

In relation to high temperature studies I would like to conclude, therefore, that inherent grain-size has a secondary relation to creep strength through its primary relation to actual grain-size; this in accordance with our conviction that it has been definitely established that actual coarse grain and actual fine grain show the lowest creep rates above and below the equi-cohesive temperature respectively.

The grain growth characteristics tell us something more about the steel than the actual grain-size we may expect after any given heat treatment. It is quite generally agreed that control of inherent grain-size is accomplished through critical dispersion of submicroscopic inclusions which induces crystallization to start at many centers and which leaves many crystal interfaces where the energy is stored at a higher level. In their paper before this meeting Davenport and Bain point out how a slight revision of our first concept of

grain-size influence should be made to include promotion of the rate of transformation by fine grain. I submit, therefore, that not only does inherent grain-size or grain growth characteristics give us a measure of the extent of the intergranular forces, but warns us that the character of these forces may be changed.

On these premises it is not illogical to suggest that the equi-cohesive temperature may change with the inherent grain-size. In the data given by White and Clark there appears to be some support for this hypothesis.

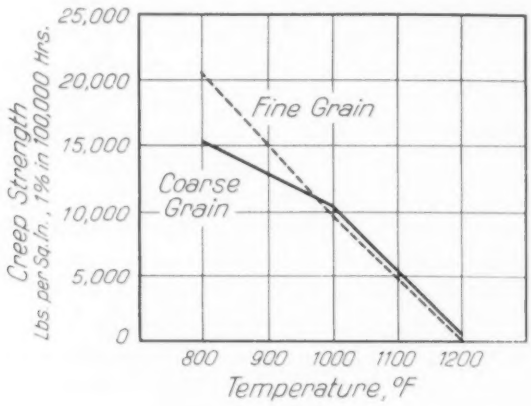


Fig. 1—Creep Strength of Fine- and Coarse-Grained Carbon-Molybdenum Steel.

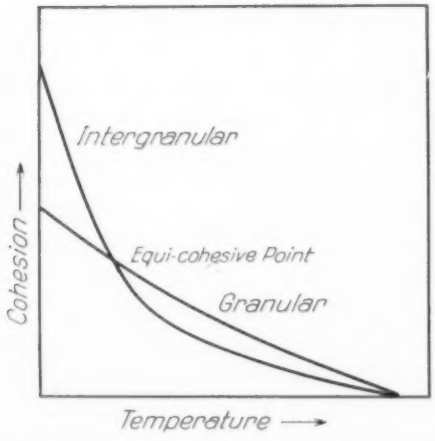


Fig. 2—Granular and Intergranular Cohesion Characteristics According to Jeffries and Archer.

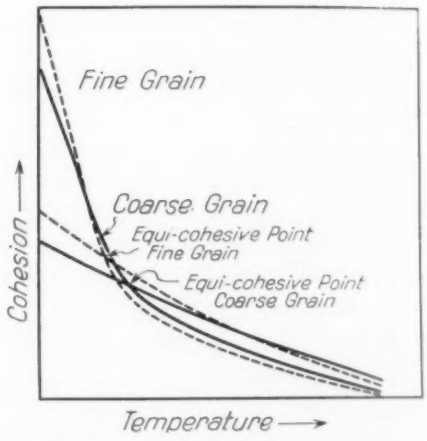


Fig. 3—Cohesion Characteristics of Fine- and Coarse-Grained Steel.

When the values given in Table II of the paper for carbon-molybdenum steel are plotted "creep strength versus temperature" the characteristic break in the creep-temperature diagram is found only in the case of the coarse-grained steel. (Fig. 1 of this discussion.)

This characteristic break occurs at the region of the equi-cohesive temperature where the creep changes from intra- to intergranular flow as in Fig. 2.

The absence of this break in the fine-grained steel creep plot suggests that

the equi-cohesive temperature for the fine-grained steel was below any of the temperatures at which creep strengths were reported. The suggestion is only entirely consistent if it can be shown how a fine-grained steel failing intergranularly can have in a restricted temperature range a higher creep strength than a coarse-grained steel failing intragranularly. To do this we must consider again how the equi-cohesive diagrams might vary for fine and coarse grain steel.

Fig. 3 shows the supposed cohesion characteristics of coarse- and fine-grained steels superimposed on the same diagram. For the fine grain we should give the intergranular cohesion a higher value at low temperatures than the coarse grain, but we should anticipate a more rapid decrease with increasing temperature. The intergranular cohesion of the fine-grained steel probably bears the same relation to the intragranular coarse grain. Such a set up places the equi-cohesive point of the fine grain steel at a lower temperature but higher stress than that of the coarse-grained steel. If creep tests are made on coarse- and fine-grained steels at a temperature slightly above the equi-cohesive temperature of the fine-grained steel and below the equi-cohesive temperature of the coarse-grained steel the result reported by the authors will be obtained. It seems probable, therefore, that the equi-cohesive temperature varies with the inherent grain-size. The authors' comments on this possibility would be valuable to us.

It is my opinion that in this paper the authors have quite firmly established their theory regarding the significance of intergranular cohesion in creep, and that through their consideration of the effect of inherent grain-size they have now essentially placed before us a full explanation of the rudiments of creep. As an outgrowth I see that we have a new method of approach in searching for compositions suited to high temperature service. The authors should be congratulated on having made another fine contribution to the technical literature on high temperature characteristics of metals.

Written Discussion: By H. W. McQuaid, research metallurgist, Republic Steel Corp., Massillon, Ohio.

The writers have shown with the steels tested that the coarse McQuaid-Ehn grain-size indicates superior creep resisting and other properties at elevated temperatures. These steels tested were in a pearlitic condition and the structures obtained would correspond closely to that obtained by the carburizing (McQuaid-Ehn) test itself. The importance of the grain-size is generally greater in the quenched condition than in the annealed condition, and it is to be regretted that samples were not tested which had been quenched and drawn to the highest temperature used in the tests.

It is also believed that the effective grain-size produced through heat treatment (actual) should have been studied by producing in the annealed condition and the heat treated condition different grain-sizes for the same tests. Varying the initial heating temperatures for quenching and annealing would have produced a change in actual grain-size which would have given to these tests greater value as far as the metallurgist interested in ferrous materials is concerned.

It would be very interesting to know the effect of annealing temperatures up to 2000 degrees Fahr. and the effect of quenching temperatures up to 1800

degrees Fahr. on the high temperature characteristics as determined by the authors.

The connection between grain-size and equi-cohesive temperature is extremely interesting. It is known that at temperatures above the Ac_3 point that a change occurs in the grain boundaries and crystal structure which results in a considerable increase in crystal size. Whether the so-called equi-cohesive temperature is due to a more rapid weakening of the crystal boundary with increase in temperature than occurs in the crystal itself, or whether an actual change takes place at a definite point in the crystal boundary, is a matter of speculation.

It is not at all improbable that at the so-called equi-cohesive temperature that there is a change in the solubility of the non-ferrous constituents of the grain boundary which changes definitely the strength of the material. The equi-cohesive temperature is in the same range as the temperatures at which nitrides are formed or go into solution, and it might be possible that the variation in the equi-cohesive temperature and in fact its existence is a function of some practice which varies the amount and condition of the nitrogen compounds present in the steel.

A. C. CHAMBERLIN:[2] The authors have effectively covered a subject of extreme interest, and it is possible that future work along this line may lead to a method for the rapid approximation of creep values on some basis other than the long time creep test. While this paper deals principally with creep and impact values, it must be remembered that other properties may influence the success of alloys used for elevated temperature service. It is believed that a study of fatigue properties at the temperature suggested by the authors would yield additional information which might minimize the importance of the superiority shown for coarse-grained steels. It is also probable that from a creep stress standpoint the manganese-molybdenum steels are somewhat superior to the straight carbon-molybdenum materials, and in this connection I would like to ask Professor White whether there was some reason for testing one steel in the annealed condition while the other was normalized and drawn. It would seem that this procedure has given an advantage to the already superior steel.

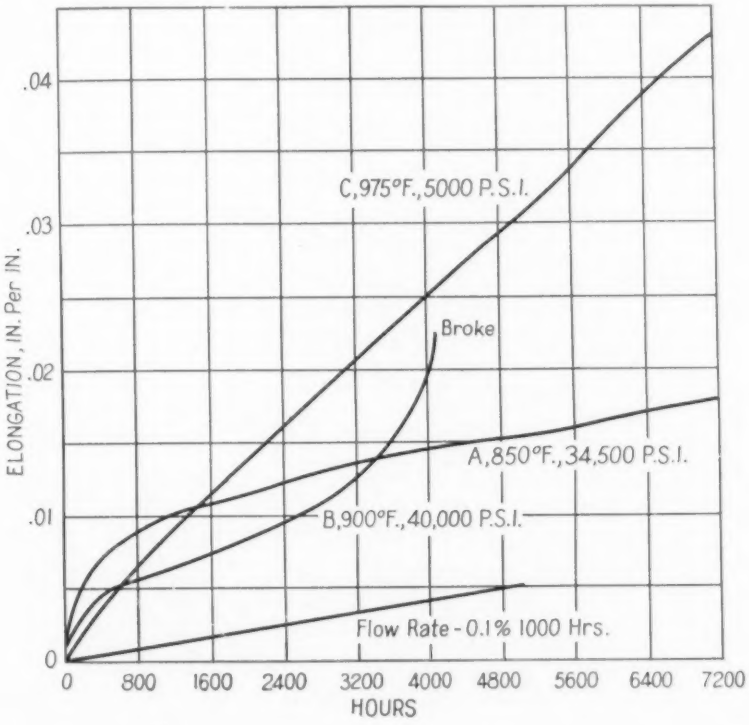
F. B. FOLEY:[3] For some time it has been common practice to plot creep curves on semi-logarithmic coordinates. The assumed straight line relationship may be a law applicable to low rates of creep, run for periods up to about 1000 hours, but I do not see how it can hold for the higher rates of creep.

The figure accompanying this discussion shows three long-time creep curves All represent pearlitic steels, but are not of the same composition. One, represented by curve C, broke after a little over 4000 hours. The other two are still running under load, having been in now for over one year and three months. Only the first 7200 hours of the runs are shown in this figure. From a consideration of these curves it is obvious that no creep rate could be estimated with any degree of accuracy from the data obtained up to 2000 hours. Nor could the future behavior have been foretold from these data. There is nothing to foretell, for example, that steel B would break, that the rate of flow of steel A would decrease, and that steel C would continue at a fairly constant rate for over

[2]Metallurgist, Bethlehem Steel Co., Bethlehem, Pa.
[3]Metallurgist, Midvale Co., Philadelphia, Pa.

a year. It will be noted that steel A, which showed the highest initial flow, has settled down to a rate of approximately 0.1 per cent in 1000 hours. This curve was obtained at the lowest temperature of the three where cold working may have been most effective.

The tests reported by Clark and White were run for 500 hours. For higher creep rates, it is doubtful if the rates determined in this short time mean anything.



It seems hardly worth while to run tests showing flow much in excess of 0.1 per cent in 1000 hours, excepting to determine the risk involved in overloading. High flow rates, when obtained in short time, appear useless in establishing stress-flow relationship.

CHAIRMAN BAIN: Mr. Foley has offered us something to think about. It seems as though we have been a little too eager to jump at the creep properties, which may not hold at all. It is interesting because his information comes from studies that certainly are not inviting to do. It is not a pleasant task to run these tests for such a long time, and we do not get as much of this information as we need.

T. W. WALKER: I should like to ask Mr. Foley if the break was transcrystalline or intercrystalline.

F. B. FOLEY: I do not know that it is possible to tell whether the fracture of specimen B was transcrystalline or intercrystalline. In a pearlitic steel it is difficult to tell. The old austenitic grain which, in hypoeutectoid steels, is defined by the ferritic network is not the true grain-size of the steel below the

critical temperature. It is, of course, the grain-size referred to in differentiating between coarse and fine grain in the McQuaid-Ehn test, but, when we are trying to determine in this case, whether the failure at temperature has occurred around the crystals or through them, the crystals we are referring to are the ferrite and cementite crystals which existed at the temperature of failure. It would be difficult to tell whether the failure occurred in the grain boundaries of the ferrite and cementite.

CHAIRMAN BAIN: I would like to point out that the grain-size that is discovered after a carburizing test at 1700 degrees Fahr. for eight hours is merely one point of a great number which determine the grain growth characteristics of the steel. It is not in itself of any more significance than the grain size after any other heating program.

Authors' Reply

The question raised by Mr. Foley as to the influence of time of test on the resulting creep characteristics is a most interesting one. Ever since the importance of creep was first realized, there has been considerable discussion as to the length of time these tests should be extended. We will all admit that the ideal procedure would be to conduct them for time periods approaching the

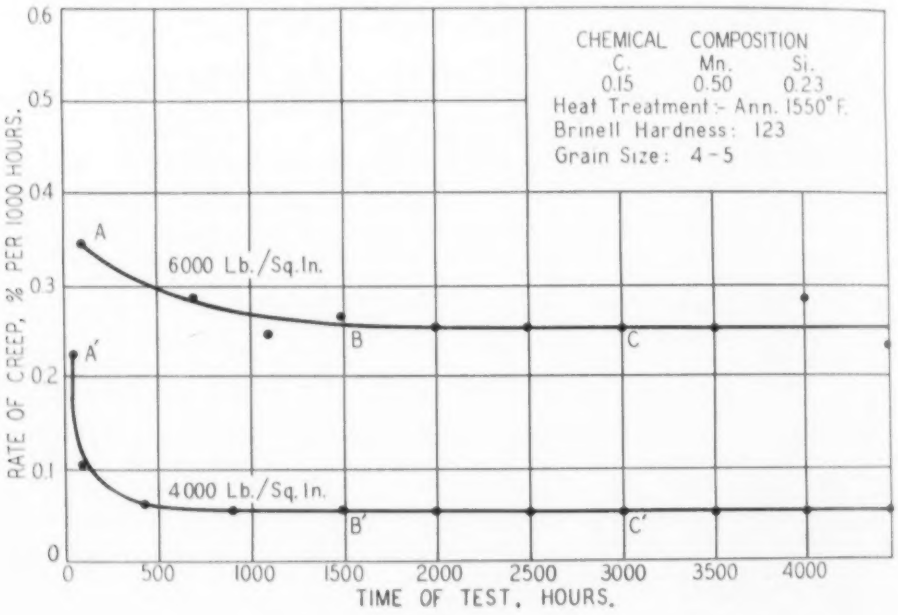


Fig. 1

commercial life desired. Such a procedure would in general be impossible, for the laboratory phase of the work would lag behind commercial progress.

We are conducting extended creep tests similar to those of Mr. Foley, and at the present time these tests have been in progress for 4500 hours. We find that the influence of time on the resulting creep characteristics can best be shown by plotting the rate of creep versus time, as in Figs. 1 and 2 of this closure. In both of these figures it is observed that the rate of creep first decreases and, in most cases it continues to do so for periods up to 1500 to 2000

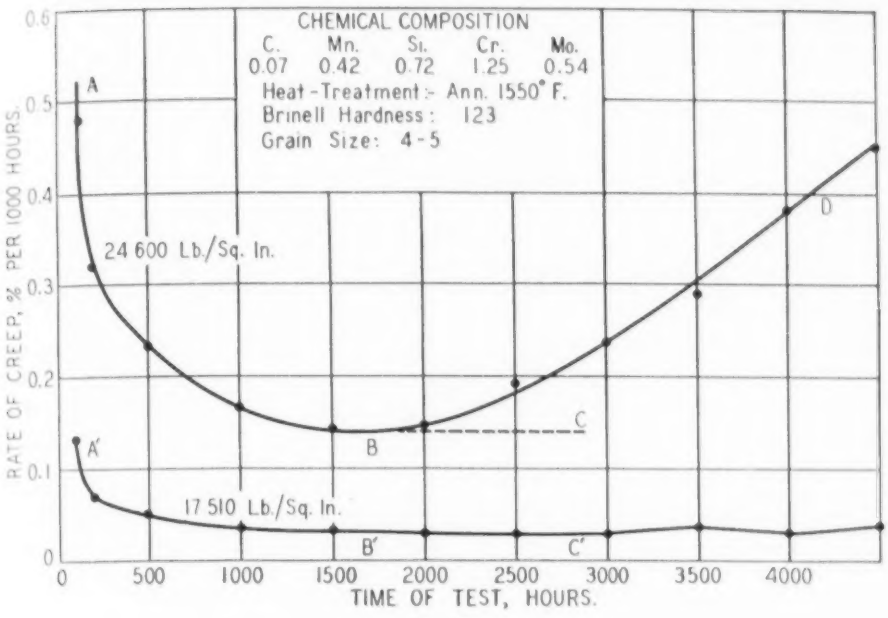


Fig. 2

hours. In one case the creep rate increases after 1750 hours, but the initial stress was purposely chosen sufficiently high to produce this condition. On the basis of these results it is our contention that tests of 500 hours give more conservative values than do tests of 1000 hours. In certain cases it is surprising as to the slight influence of extended time periods. For example, with the plain carbon steel at 1000 degrees Fahr., Fig. 1, the stress for a creep rate of 1 per cent per 10,000 hours would be 4500 pounds on the basis of the 500 hour tests, and 4700 pounds on the basis of the 4500 hour tests, a difference of only 200 pounds for a difference in testing time of 4000 hours.

In regard to the question of Mr. Chamberlin as to why one steel was tested in the annealed condition and the other in the normalized and drawn, that goes back to the type service for which these steels are intended. It is realized that the 0.50 molybdenum steel, when used in oil cracking work, may approach temperatures of 1200 degrees Fahr. or higher, and, therefore, any normalized and drawing treatment would be entirely erased. The manganese-molybdenum alloy, Steel MM9, is especially intended for operating temperatures up to 1000 to 1100 degrees Fahr.

While we were interested in showing in this paper the influence of grain-size on certain high temperature characteristics, we also were interested in making all conditions such that the results obtained could be commercially used. It was for this reason that the quenched and drawn treatment, as suggested by Mr. McQuaid, was not employed. Results from other tests, however, indicate that at these higher temperatures a quenched and drawn structure generally produces creep characteristics inferior to those produced by either an annealing or a normalizing and drawing treatment. Other tests also indicate that the creep resistance at the higher temperatures can be increased by increasing the annealing or normalizing temperature. This probably is due in part at least to the larger actual grain-size which results.

The authors agree with Mr. Newell in that creep characteristics alone should not determine the selection of material for high temperature service. Other physical properties such as ductility, structural stability and freedom from embrittlement may be of equal or even greater importance.

The authors greatly appreciate the points raised by Mr. Herzig and believe his suggestion as to the dependence of the equi-cohesive temperature on inherent grain-size to be a good one. Creep is, in general, a result of two factors, strain hardening and recrystallization. Since the inherent grain-size is a measure of a metals susceptibility to grain growth at sufficiently elevated temperature, it is logical to believe it may also influence the recrystallization tendency, since the two phenomena, grain growth and recrystallization, are more or less related. Any factor which influences the recrystallization tendency will, of course, influence the equi-cohesive temperature, which may be defined as the lowest temperature of recrystallization.

THE IMPORTANCE OF GRAIN-SIZE OF SHEET STEEL FOR DEEP DRAWING

By Reid L. Kenyon

Abstract

The physical properties of steel sheets for deep drawing are affected by several variables. Although grain-size is one of these, composition, structure and thermal and mechanical treatment are also influential and at times may have a greater effect. Grain-size affects the smoothness of the surface of the drawn sheet and if this is important the grains must not be too large. Coarse-grained material is somewhat softer and weaker but also has lower ductility than finer-grained sheets. Manufacturers of deep drawing sheets are aware of the importance of the grain-size of their product and are able to control this within limits that are closer than the differences that can be observed in the performance of the material. The refinement of the grains in cold-rolled strip is due to recrystallization during low temperature annealing (below A_3). Experience in producing steel sheets for deep drawing has shown that physical tests together with microscopic examination furnish satisfactory means for controlling the drawing quality, assuming, of course, that the other variables such as composition, and thermal and mechanical treatment are also controlled.

THE requirements of sheets for deep drawing are constantly becoming more exacting. In addition to larger sizes and lighter gages, users are demanding better surface and improved physical properties. There are several variables that have an important influence on the physical properties, among which may be mentioned composition, structure, and thermal and mechanical treatment. These are more or less inter-related, but attention will be confined, in this discussion, to the effect of structure, and even be limited to one feature of the structure—grain-size. It should be borne in mind, however, that these other variables are present in all cases and at times may have a greater effect than the one now under consideration. It is true, though, that grain-size does have a very important role to play

A paper presented as part of the Grain-Size Symposium of the Sixteenth Annual Convention of the Society held in New York City the week of October 1, 1934. The author, Reid L. Kenyon, a member of the Society, is Supervising Research Engineer, American Rolling Mill Co., Middletown, Ohio. Manuscript received June 25, 1934.

in the development of superior deep drawing sheets because it affects both the physical properties and the surface of the finished product.

Effect of Grain-Size on Surface

The grain-size may have a marked effect on the smoothness of the surface of the drawn part. If a metal is deformed beyond the

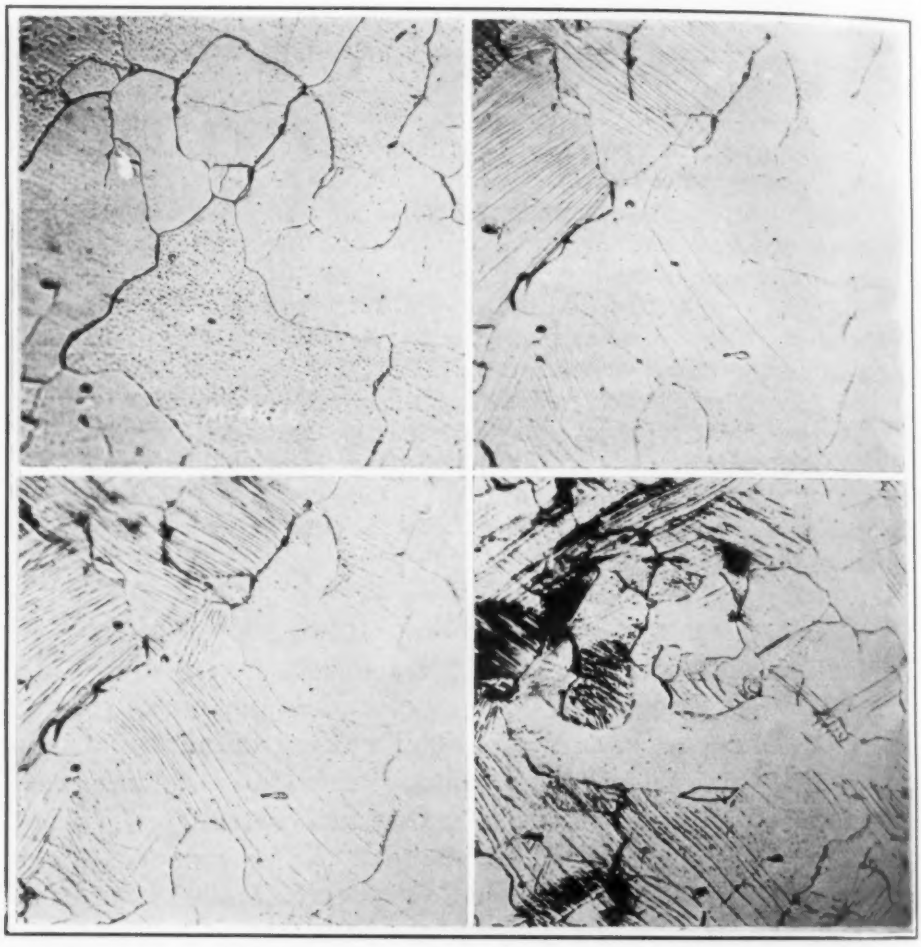


Fig. 1—Successive Stages in the Formation of Slip-bands in Armco Ingot Iron. × 100.

elastic limit, the orderly arrangement of the structure is altered. There are two types of internal movement that occur during plastic deformation—slip and rotation. These probably take place simultaneously. Slip may be detected by the presence of "slip-bands" on polished and etched micro-specimens. Successive stages in the

formation of these slip-bands on a piece of ARMCO ingot iron are shown in Fig. 1. The piece was polished and etched and the first photograph taken to show the appearance before deformation. The specimen was then compressed enough to cause a slight amount of plastic deformation and the second photomicrograph shows the appearance of numerous dark lines which are recognized to be the

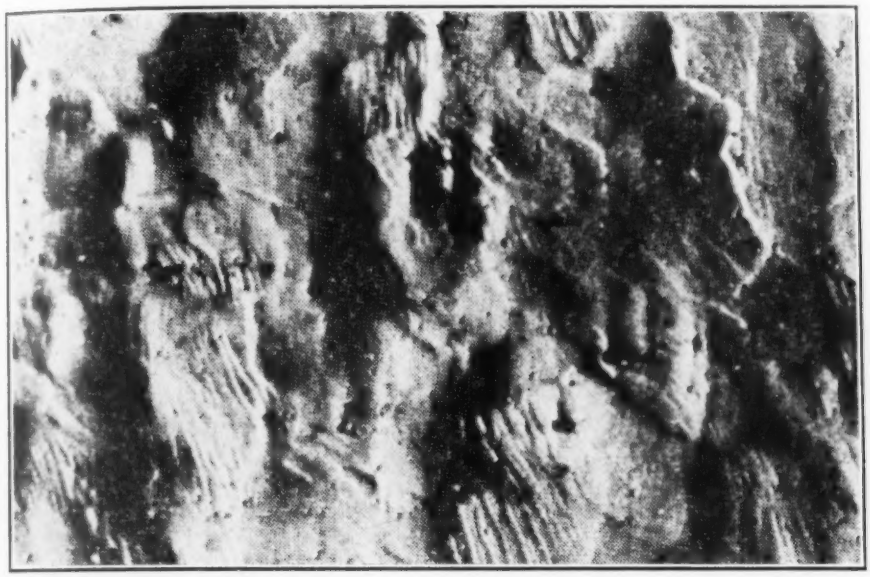


Fig. 2—Roughening of Polished Steel Surface Due to Severe Deformation (Nadai).

traces of slip planes on the plane of polish. The effect of further deformation is to increase the number of these slip-bands, as is shown in the third photomicrograph. If the deformation is carried still further, secondary sets of slip-bands cross the first set. As the slip proceeds, end interference from the adjacent grains increases and the rotation of the grains proceeds far enough to produce a visible effect on a polished surface. Such a surface, deformed severely enough to cause roughening, is shown in Fig. 2. The various grains are inclined at different angles showing that there must have been rotation to destroy the originally plane surface. If the grains are fairly large, the variation in their inclinations can be seen without magnification, but this "upheaval" effect is present even on fine-grained material, although in such cases, it can be observed only under a considerable magnification and then the original surface must have been highly polished. The surfaces ordinarily found on commercial

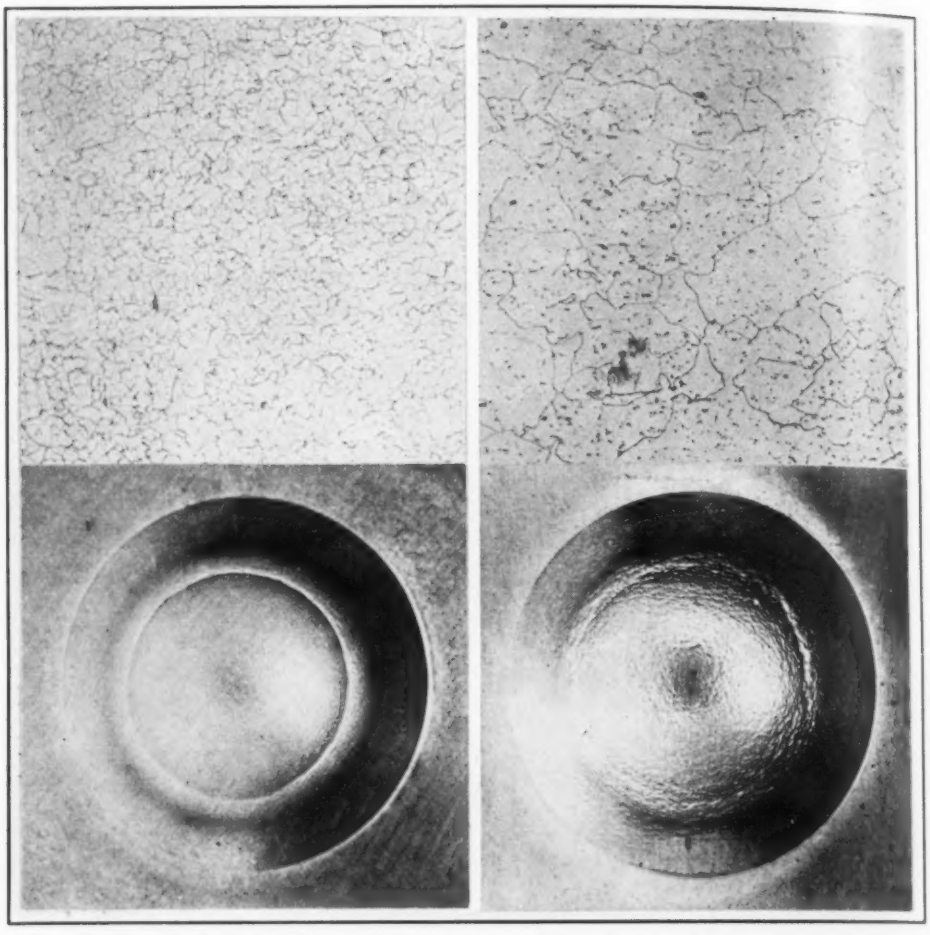


Fig. 3—Erichsen Draws and Microstructure (100 ×) of Fine-Grained Normalized (Left) and Coarse-Grained (Right) Box Annealed Drawing Sheets.

deep drawing sheets and strip do not show any coarsening that can be detected by the unaided eye if the grains are sufficiently small.

The "cupping" tests such as Olsen and Erichsen are useful in showing the character of the surface of the sheet after it has received a severe draw. Erichsen draws made on sheets having different grain-size are shown in Fig. 3. Above each specimen is the photomicrograph of its grain structure. This illustrates the range of grain-sizes considered in this discussion and shows the condition of the surface when sheets exhibiting such structures are badly deformed. The coarsest grains are similar to those observed in overheated box annealed material. This condition may also result if the material has been cold-worked a critical amount so that the subsequent box annealing causes abnormal grain growth. The fine grains

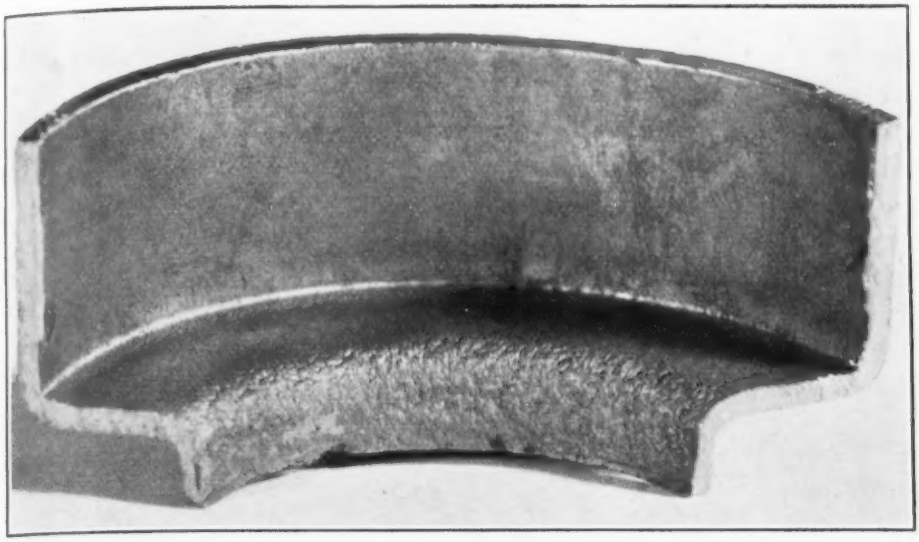


Fig. 4—Roughened Surface of Sharp Corner on an Automobile Hub Cap.

represent normalized material. The roughened surface observed in the case of the coarser-grained specimen is quite unsuited for many finishes that are applied to drawn parts; for example, automobile body parts would appear rough through the lacquer finish. This also applies to pieces such as hub caps and radiator shells which are electroplated.

Sheets for deep drawing have to meet various requirements due to the wide range of designs and shapes fabricated and due to differences in the design of the dies. These variations result in more or less severe working due to the manner in which the metal must flow to make the draw. In many cases, certain portions of the part must undergo large local elongation and this is likely to develop a rough surface if the material is coarse-grained. This condition is noticeable around beads and corners. A portion of a hub cap which has developed a very rough surface around a sharp corner is shown in Fig. 4. This condition may be remedied in many cases by grinding, filing, or hand polishing, but in instances like this one, such operations would be very difficult if not entirely impracticable. In any event, this would be an item of added cost which is, of course, highly undesirable.

There are certain parts in which the curvature may be very much less than that required around corners and beads, but in which there is considerable general elongation. Here again the surface of coarse-grained material may roughen. A shroud part drawn from a coarse-

grained sheet is shown in Fig. 5. There is only a moderate curvature to the portion of this draw shown in the illustration, but the material was stretched enough to develop a very rough surface.

There are still other parts that do not undergo severe deformation, either local or general, and in such cases a coarse-grained structure is not detrimental. As a matter of fact, the extra dead

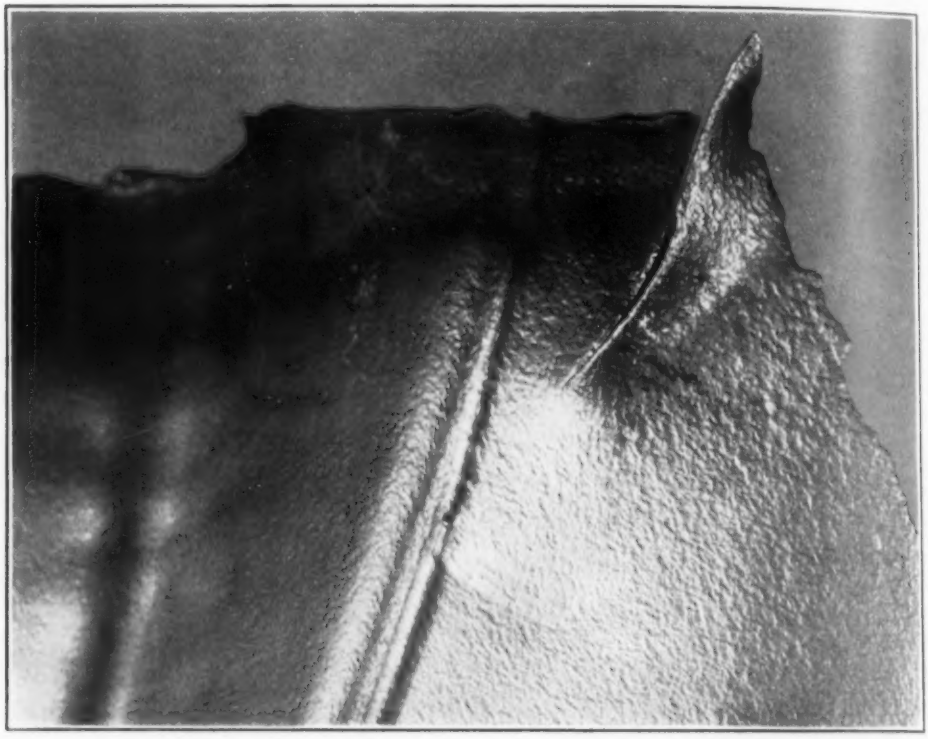


Fig. 5—Rough Surface Due to Coarse Grains on Shroud of Automobile Body.

softness—lack of springiness—which accompanies larger grain sizes, may be quite desirable in order to insure that the metal takes a definite shape and does not spring away from the die or form, around which it may be bent. Such an application requires a very soft sheet. Extremely fine-grained sheets are found to be somewhat harder than coarse-grained ones. A grain size must, therefore, be produced that will satisfy the requirements of both softness of the sheet and the desired degree of smoothness of the surface after drawing.

Effect of Grain-Size on Physical Properties

The grain-size has an important influence also on physical

properties. Mention has been made of the mechanism of plastic deformation. The first essential is that slip can occur along certain planes which experiment has shown correspond to the orientation of the space lattice of the individual grains. Since the continuity of the lattice and therefore of the slip planes is broken at the grain boundaries, it is apparent that the number of these grains boundaries must

Fig. 6—Tensile Tests on Coarse-Grained Aluminum Showing Stiffening Effect of Grain Boundaries. (Sachs)

have a marked effect on the ease with which slip can take place. The mutual supporting of the adjacent grains is shown by the deformation of tensile specimens that are made up of only a few grains. Such tensile specimens of aluminum are shown in Fig. 6. The grain boundaries have caused "humps" on the surface due to their resistance to deformation as the specimen was stretched in the tensile test.

The supporting effect of these grain boundaries extends a considerable distance within the grains. A decrease in the distance between these boundaries would, therefore, increase the strengthening effect even if there remained only one grain across the entire section. If the number of grains on the cross section is likewise increased, the supporting action is still greater. With relatively large grains the number of them on the cross section is of importance, but as the number increases, the absolute size is more important. The shape of the section may also be of influence if one dimension is quite small.

Although the effect of grain-size on mechanical properties is generally acknowledged, relatively few data have been published. This lack of information is partly due, no doubt, to the difficulty of completely isolating this variable from others which must be present in varying degrees due to the methods employed in obtaining the different grain-sizes in the material under study. For example, in low carbon steel the grain-size can be varied either by cold rolling different amounts and annealing at temperatures in the recrystallizing range and below the A_3 point, or by heating above the A_3 point for various

lengths of time and cooling at various rates. Although a variation of these factors will produce a variation in grain-size, there may also be a difference in the form of other constituents such as carbides and in the amount of unrelieved strain from the previous rolling operation.

Bassett and Davis (1)[1] have published tests on alpha brass showing how the Brinell hardness varies with grain-size. The results

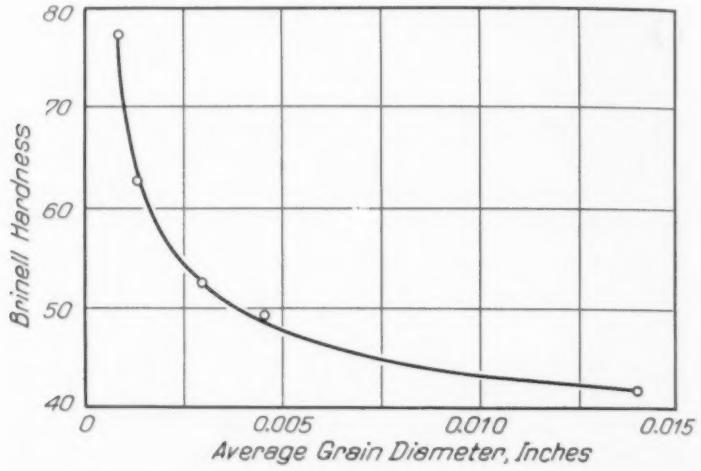


Fig. 7—Relationship Between Grain-size and Brinell Hardness of Alpha Brass. (Bassett and Davis.)

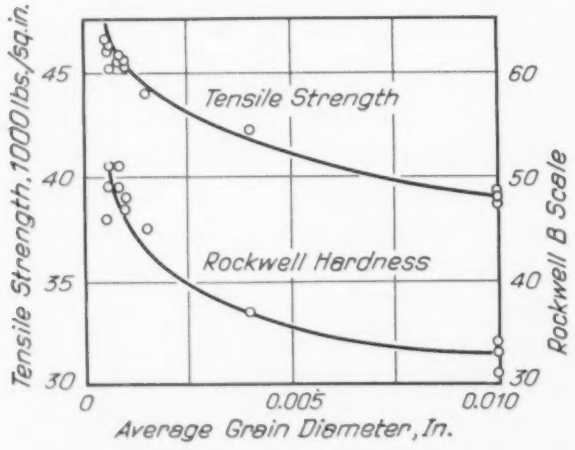


Fig. 8—Relationship Between Grain-size and Rockwell Hardness and Tensile Strength of Mild Steel Sheets.

of their tests are shown in Fig. 7. Small changes in grain size have a more pronounced effect when the grains are small. This is influ-

[1]The figures appearing in parentheses refer to the bibliography appended to this paper.

enced by the size of the grains relative to the cross section. The effect of grain-size on the tensile strength and Rockwell hardness of low carbon steel sheets for deep drawing is shown in Fig. 8. These specimens were prepared by cold rolling the sheets different amounts and then annealing to get different grain-sizes. The other factors, such as form of the carbon and rolling stresses, were probably fairly well eliminated because the carbon appeared in the form of cementite islands in all samples and the annealing temperature was high enough to relieve most of the rolling stresses. The finer-grained material is harder and stronger than the coarse-grained material.

Although the grain-size of ferrous sheets may vary over the entire range indicated in Fig. 3, a large proportion of the deep drawing material now produced in this country is normalized and has relatively small-sized grains compared with box annealed sheets. Differences in the physical properties of the large-grained box annealed material and the fine-grained normalized sheets are much greater than the differences that occur between the so-called large-grained and fine-grained normalized material. The range of grain-sizes found in commercial normalized sheets for deep drawing is from the finest box-annealed structure to the finest normalized structure. Sheet manufacturers have found that this range of grain-sizes gives satisfactory performance in deep drawing work. Experienced producers are able to manufacture sheets of the proper structure and physical properties for the various types of deep drawing jobs and this depends to a large extent on the design of the part, the design of the dies for making it, and the practice followed in the drawing shop. Some users prefer a finer-grained sheet in order to insure against surface roughening, while others prefer a softer sheet with a slight risk as to surface. For the above reasons a noticeable difference may be observed in the performance of material of different grain-size from different mills. Either material would work properly with suitable die practice.

Many drawn parts having no sharp radii could be satisfactorily made from sheets having a considerable difference in grain-size. In such cases the differences found in commercial normalized deep drawing sheets would have no important effect on the performance. Even where die shop practice and press capacities make it advisable to supply material of a certain grain-size, the sheet manufacturer is able to control this much more closely than can be detected by drawing performance.

There is another important point in connection with the relationship between grain-size and physical properties. In order to prevent stretcher straining, the sheets must necessarily be given a final cold rolling, the amount depending on the particular type of draw they are to be used for. This also affects the physical properties so that the user is unable, from any tests he may make, to correlate grain-size with performance in the die shop. The sheets as he gets them have other variables imposed upon the one in which he may be interested. Sheets of the same grain-size may be made to have widely different properties and conversely sheets having similar physical properties may be made with different grain-size, depending on the operations performed on them. For example, a coarse-grained sheet, which would naturally be rather soft after annealing, may be cold-rolled enough to make it as hard as a fine-grained sheet which has had very little or no cold rolling. The two sheets will have vastly different drawing properties.

Effect of Heat Treatment on Grain Size

The metallography and heat treatment of low carbon steel has been adequately discussed in various books, such as those written by Professor Sauveur, and by Jeffries and Archer, but it may be well to summarize the main points briefly. As the steel is heated, it passes successively through the lower critical points, Ac_1 and Ac_2, but any ferrite remaining, due to slow diffusion rates, retains the same lattice form until it passes through the A_3 point and then changes from the alpha, body-centered form to the gamma, face-centered form. The reverse process occurs on cooling. This complete recrystallization is due to the allotropic change from the alpha to the gamma form and back to the alpha. To quote from Jeffries and Archer (2)—"When pure iron is heated above 900 degrees Cent. (1652 degrees Fahr.) so as to cause the A_3 allotropic change to take place, a complete change in its crystalline structure occurs. The body-centered cubic lattice of the alpha iron changes to the face-centered cubic lattice of gamma iron. The grains of gamma iron form from many centers or nuclei, which are usually at the grain boundaries of the alpha iron. The grain boundaries of the gamma iron do not coincide with old grain boundaries of the alpha iron. In other words, the transformation of alpha to gamma iron not only represents a change in space lattice, but also the entire disappearance of the alpha grains

and the birth of new gamma iron grains, whose size and shape are determined by the ordinary conditions affecting grain growth. The A_3 transformation point is in a temperature range of grain growth for both alpha and gamma iron.

"When gamma iron is cooled through the A_3 point so as to form alpha iron, an entirely new set of alpha iron grains is formed. * * * * The new alpha grains begin to form at the grain boundaries of the gamma iron. The size and shape of the new alpha grains will depend largely on the conditions obtained during and after the transformation. Slow cooling allows time for growth of the alpha grains and favors the formation of relatively few nuclei. Rapid cooling, as, for example, water-quenching from above A_3, promotes the formation of more nuclei and retards grain growth after the transformation is complete."

Although extremely rapid cooling, such as by water or oil-quenching from just above A_3, would produce fine grains, there would also be undesirable effects, especially in sheets, due to the effect on the carbon which is in solution at the temperature to which the material has been heated. Such a rapid cooling rate would hold the carbon in the form of martensite or sorbite and so produce a very hard sheet of relatively low ductility. Furthermore, the rapid cooling would leave many internal stresses and cause warping. This treatment would be entirely impractical on sheets and it is fortunate, therefore, that it is not necessary to resort to such rapid cooling rates to obtain the desired grain structure.

When carbon steels are heated above the A_3 temperature and allowed to cool in still air, the treatment is called normalizing and this is often thought to be a perfectly uniform treatment that assures comparable structure and properties for all sorts of pieces of the same composition. The actual cooling rate, however, depends on the size of the piece; larger thicker pieces cool more slowly than smaller thinner ones. The number of sheets in a pack, the design of the hood, and type of conveyor also affect the cooling rate. It is, therefore, incorrect to assume that normalizing will produce identical structures and properties in all steels regardless of the dimensions of the piece treated.

The principal factors of heat treatment which affect the grain-size of low carbon steel sheets for deep drawing are the time and temperature of annealing and the cooling rate. These are the variables that the producer can control to get the grain-size that, in his

judgment, is best suited for the particular job. All sorts of automatic control equipment have been developed for maintaining close limits on all of these variables so that the degree of uniformity found in present-day deep drawing sheets is really remarkable and would have been considered impossible a few years ago.

The general principles of box annealing and normalizing are well and generally understood, but may be briefly mentioned. In the case of box annealing, time and temperature are the most important variables. The higher the holding temperature and the longer the time at this temperature, the larger the grain structure. Because all of this annealing is done below the A_3 temperature, the recrystallization that occurs is due to the relief of internal stresses that were caused by the rolling operation. In the case of normalizing, all three variables have an important effect and are carefully controlled in the manufacture of deep drawing sheets. The material is heated above the A_3 temperature for a short period of time and then allowed to cool either directly in the air or at a somewhat retarded rate by means of a hood over the exit conveyor from the furnace. This is a matter of furnace design and has been described in various articles, among which may be mentioned those by Whitefield in the *Journal* of the British Iron and Steel Institute in 1930, Lawrence in the TRANSACTIONS of the American Society for Steel Treating in 1931, and Davis in *Fuels and Furnaces* for March, 1932. As mentioned before, the more rapid the cooling rate, the smaller the grains. It is impossible, however, in commercial normalizing furnaces, to obtain grains as large as those often found in over box annealed material.

Another class of material used for deep drawing work is cold-rolled strip. This material is annealed at temperatures below the A_3 temperature. The recrystallization in this case is due not to an allotropic transformation, but to the relief of internal stresses produced by cold working. This is a very interesting phenomenon and has been the subject of a great amount of experimental work. Abnormal grain growth in low carbon steel annealed below A_3 was noticed by Stead (3) in 1898. In 1912 Sauveur experimented with some 0.05 per cent carbon steel which he heated to 650 degrees Cent. (1200 degrees Fahr.) for 7 hours after making a Brinell impression. Fig. 9 shows the grain growth he found to take place under the impression. Attention is called to the fact that the coarsest grains do not occur where the metal is worked the most severely, near the

impression, but at a rather deep point under the impression. Furthermore, there is a very sharp line of demarcation between the coarse grains and the fine grains of the undisturbed portions of the piece. This indicates that some intermediate and critical amount of internal stress is necessary to cause abnormal grain growth.

As a result of this and other similar experiments, Sauveur (4)

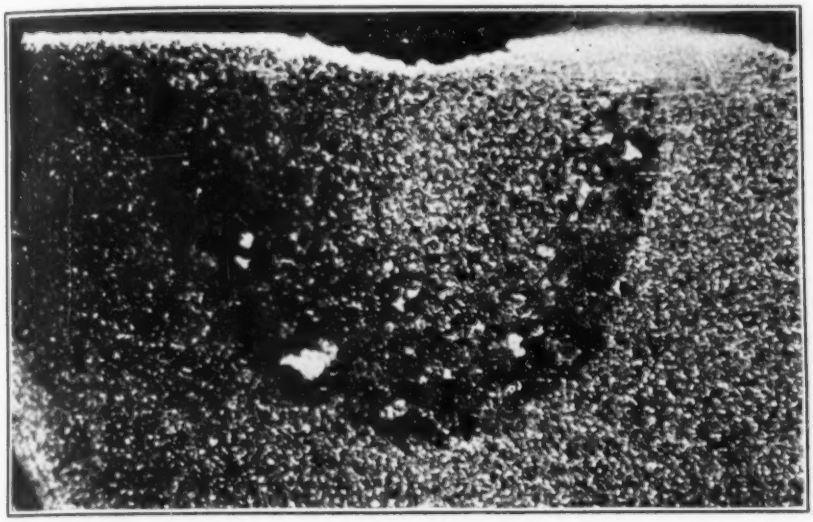


Fig. 9—Grain Growth Under Brinell Impression Due to Recrystallization Caused by Annealing at 650 Degrees Cent. for 7 Hours. (Sauveur)

concluded that a certain stress is necessary to cause grain growth of the ferrite grains on annealing below the A_3 temperature and that they will not grow if that stress has been exceeded. In 1913 Robin (5) observed the "Sauveur phenomenon" on nonferrous metals and Ruder (6) on silicon steel. In 1914 Chappell (7) performed extensive experiments on recrystallization. Fig. 10 shows one of his tapered tensile test specimens which shows very conclusively that the abnormal grain growth occurs only within a certain range of stress. He also studied the effect at various temperatures and found that less stress was necessary for abnormal grain growth when the annealing temperature was higher. A quite complete study of recrystallization due to annealing after cold working, was made by Jeffries in 1916 and published in numerous places (8) together with discussions and contributions by many other workers. Jeffries and Archer devote an entire chapter to this subject in their book "Science of Metals." One of the most important points to remember in this connection is that there is a critical range of cold working which

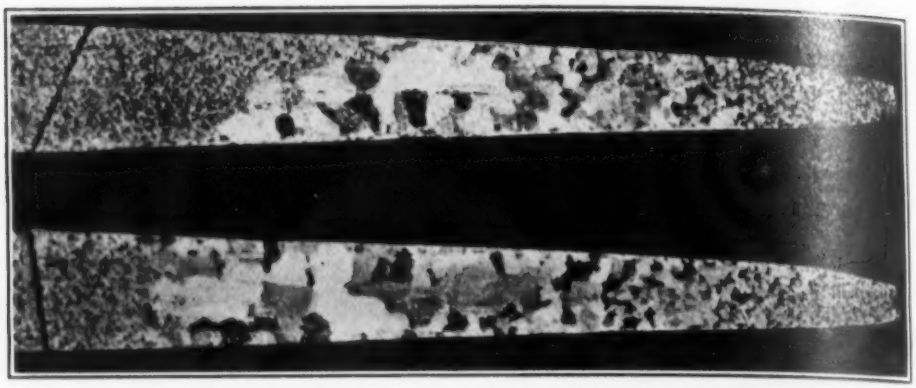


Fig. 10—Grain Growth in Tapered Tensile Specimens Due to Recrystallization after Cold Working. (Chappell)

will result in abnormal grain growth upon annealing in a range of temperature which corresponds to that usually used in box annealing of low carbon steels. This difficulty is overcome in practice by cold working an amount in excess of the critical range. For example, it has been found that reductions of about 7 to 15 per cent will produce coarse grains upon annealing, depending somewhat on the annealing temperature. In practice the strip is reduced 30 per cent or more before annealing and this avoids all possibility of critical grain growth and actually produces a satisfactory fine grain structure.

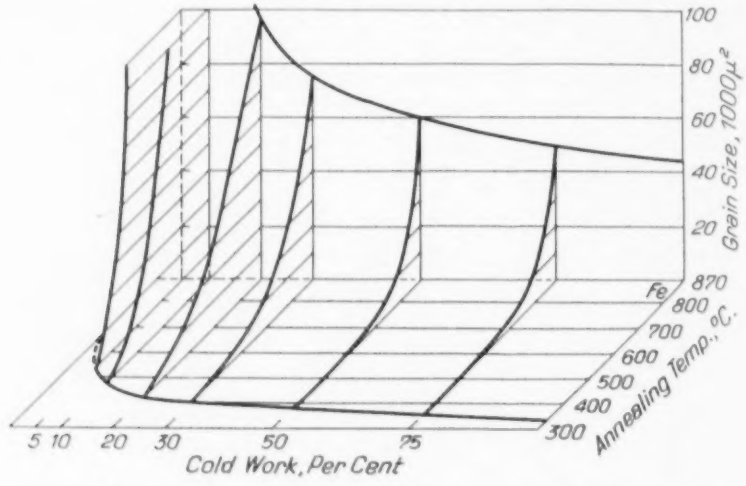


Fig. 11—Recrystallization Diagram for Electrolytic Iron. (Oberhoffer)

The relationship between amount of cold working, annealing temperature and resultant grain size is shown in Fig. 11, as de-

termined by Oberhoffer (9) for pure iron. This may be taken as representative of the general relationship between these variables for low carbon steel as well. Although this illustration does not show the effect of other variables such as original grain-size and the time of holding at temperature, it does give a good idea of the influence

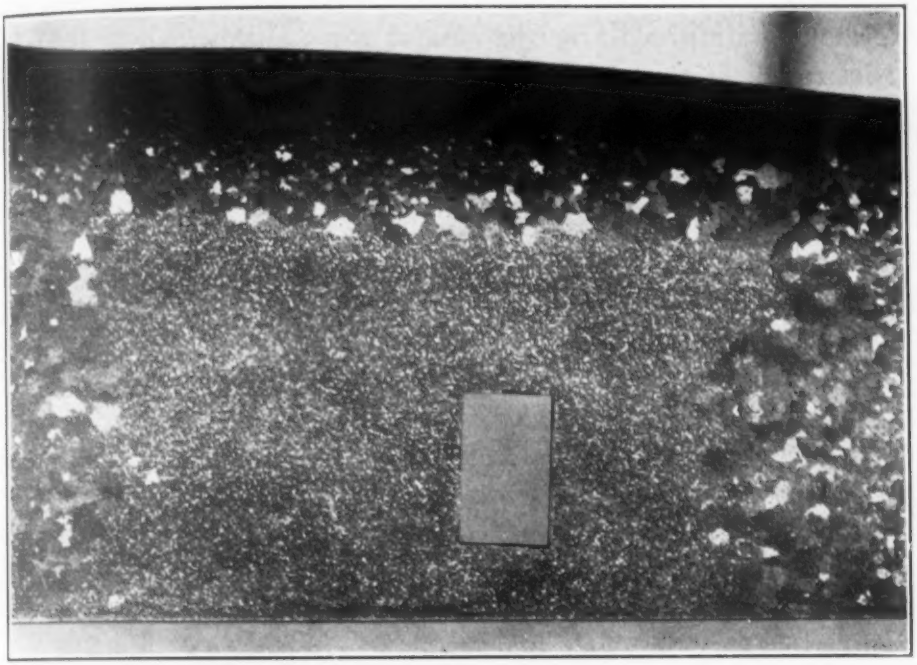


Fig. 12—Grain Growth in Drawn Shape Due to Recrystallization after Cold Working.

of the variables considered. The control of grain-size in cold-reduced box-annealed material is therefore based on the amount of cold rolling in addition to the time and temperature of annealing. The remarks that have already been made concerning the importance of grain-size on drawing performance apply to this class of material as well.

There is another lesson that may be learned from our experience with recrystallization of cold-worked materials. It is sometimes necessary to make an intermediate anneal between drawing operations and the temperatures of furnaces available for such annealing are often in the range that may cause critical grain growth. In a drawn shape various portions receive different amounts of cold work and it is almost certain that some parts may have been deformed a critical amount, that is, between 7 and 15 per cent. These will show abnormal grain growth after a low temperature anneal.

All of the effort which the sheet manufacturer has expended to get a good structure will then be lost, for the resultant coarse grain will cause trouble in subsequent drawing operations. The only solution to this difficulty is to design the drawing operations so that intermediate anneals are not necessary or else to anneal either below the critical grain growth temperature range or to normalize by heating above the A_3 point and air cooling. Fig. 12 shows the enormous grain growth that may occur when a drawn shape is given an intermediate anneal in the temperature range that causes critical grain growth.

Summary

In conclusion it may be well to summarize what has been said. Grain size is only one of several important variables in sheets for deep drawing purposes, and it alone does not determine the drawing performance. Differences between the grain-size of box-annealed and normalized sheets have a profound influence on the drawing quality as well as the surface of the finished part. Over this range, the finer-grained material is somewhat harder, but is more ductile and produces a much smoother surface after severe deformation. Commercial normalized sheets today show a variation in grain-size over a range which is much narrower than that just mentioned. To a large extent this range is determined by the experience of sheet manufacturers in producing sheets to make specific jobs. The design of the dies and the drawing shop practice enter into this and, although certain users find a given type of sheet best suited to their needs, it does not mean that another type cannot be made to do the same job. Sheet manufacturers recognize the importance of controlling the grain-size of sheets for deep drawing and are able to regulate this within limits that are closer than the differences that can be observed in the performance of the material. Recrystallization during annealing after cold working is responsible for the grain refinement in cold-rolled strip and also for the coarsening of the grains in some deep drawn parts that are subsequently annealed in the critical temperature range. The latter difficulty can best be overcome by normalizing. Experience in producing steel sheets for deep drawing has shown that physical test data together with a knowledge of the grain-size are satisfactory indications of drawing quality and a control of these properties makes it possible to produce material that meets the requirements of present deep drawing practice.

This presupposes that the other variables such as composition, and thermal and mechanical treatment are also controlled.

References

1. W. H. Bassett and C. H. Davis, "A Comparison of Grain Size Measurements and Brinell Hardness of Cartridge Brass," *Transactions*, American Institute of Mining and Metallurgical Engineers, Vol. 60, p. 428.
2. Jeffries and Archer, "Science of Metals," McGraw-Hill Co., 1924, p. 414.
3. Stead, "Crystalline Structure of Iron and Steel," *Journal* of the Iron and Steel Institute, 1898, No. 1, p. 145-205; "Brittleness Produced in Soft Steel by Annealing," ibid., 1898, No. II, p. 137-184.
4. Sauveur, "Note on the Crystalline Growth of Ferrite Below its Thermal Critical Range," *Proceedings* of the International Association for Testing Materials, Sixth Congress, New York, 1912, Vol. ii. II, 6.
5. Robin, "Recherches sur le Developpement des Grains des Metaux par Recuit apres Ecrouissage," *Revue de Metallurgie*, 1913, vol. x, p. 722-757; Developpement des Grains de Recuit dans les Alliages, vol. x, p. 758-768, ibid., 1913.
6. W. E. Ruder, "Grain Growth of Silicon Steel," *Transactions*, American Society of Mechanical Engineers, 1913, Vol. xlvii, p. 569-583.
7. Chappell, "Recrystallization of Deformed Iron," *Journal*, Iron and Steel Institute, 1914, No. I, p. 460-498.
8. Jeffries, "Grain Size Measurement in Metals," Chemical and Metallurgical Engineering, Vol. 18, 1918, p. 185. "Grain Growth Phenomena in Metals," *Transactions*, American Institute of Mining and Metallurgical Engineers, Vol. 56, 1916, p. 571-581. "Grain Growth in Metals," *Journal*, Institute of Metals, No. 2, 1918, p. 109. "Grain Size Inheritance in Iron and Carbon Steel," *Transactions*, American Institute of Mining and Metallurgical Engineers, Vol. 58, 1918, p. 669-695.
9. Oberhoffer and Oertel, *Stahl und Eisen*, Vol. 39, 1919, p. 106; Vol. 44, 1924, p. 560.

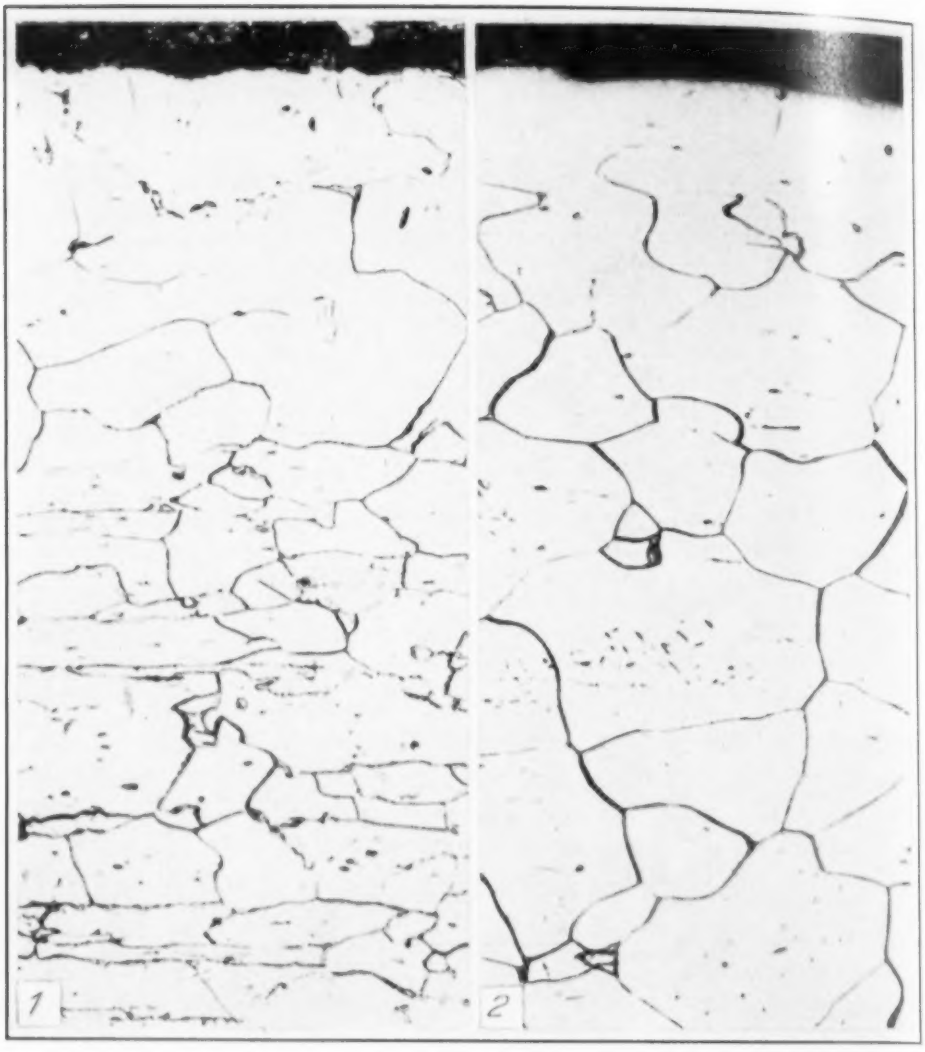
DISCUSSION

Written Discussion: By G. R. Brophy, metallurgist, General Electric Research Laboratories, Schenectady, N. Y.

The ideal situation is dealt with by Mr. Kenyon in his paper, and the efforts of the steel manufacturers toward this end are commendable. It seems, however, that there is one limitation placed upon the manufacturer of deep drawing steel which will keep him far short of the ideal. That is the nature of the material itself.

This steel is of the so-called rimming type, which of necessity varies in its nature from the surface to the center of the ingot. This natural difference carries through to the finished sheet, and as a result sheets are found usually with dual structure.

The surface material of any sheet is inherently coarse-grained, while the core is inherently fine-grained. Those sheets, or portions of sheets, which are reduced wholly from rim material, will be naturally coarse-grained. As a result the response to mechanical, and especially thermal treatments, will vary accordingly.



As an illustration, two sheets of automobile body steel from the same sheet bar analyzing 0.03 per cent carbon and 0.28 per cent manganese were given the same mechanical treatment and then annealed at the same time at 740 degrees Cent. (1365 degrees Fahr.). The resulting structures are shown in Figs. 1 and 2.

After a normalizing treatment from 930 degrees Cent., the structures shown in Figs. 3 and 4 resulted.

The hardness and drawability of these materials vary beyond the working range for this class of steel, and therefore the question arises how is it possible for the manufacturer to so closely control his grain-sizes?

Written Discussion: By N. A. Ziegler, Power Use Research Department, the West Penn Electric Co., Pittsburgh.

The grain growth in cold-worked low carbon steels and carbonless iron at temperatures below the transformation range, as pointed out by the author, is a

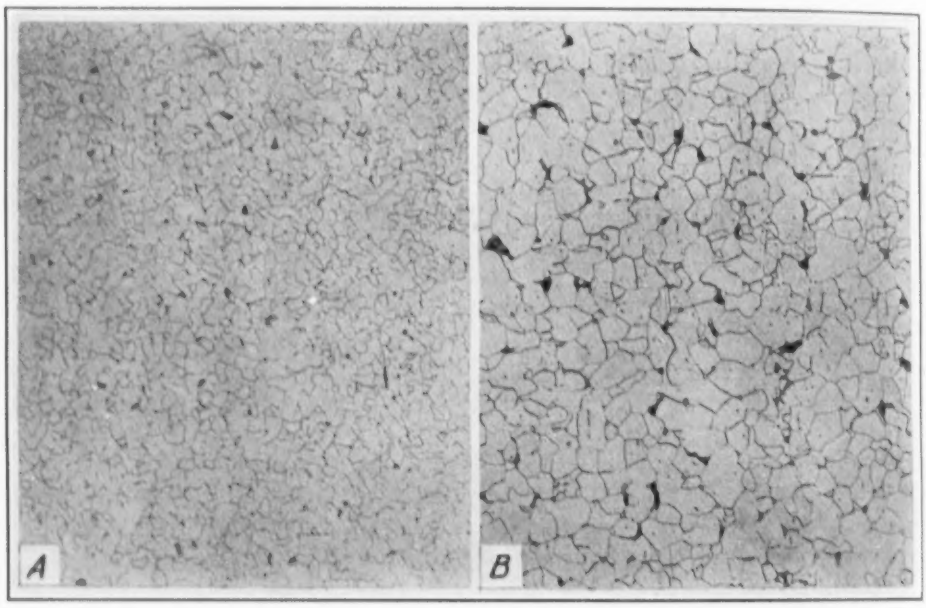
very interesting phenomenon. Several years ago I had a chance to tackle it from a different angle then described in this paper, namely, in attempting to produce single crystals of iron. Part of this work was published elsewhere (Proceedings of American Institute of Mining and Metallurgical Engineers, Iron and Steel Institute, 1930, p. 209). In preparation of single crystals of pure iron "Edwards and Pfeil" method, which consists in straining iron samples above the elastic limit and subsequent heating at temperatures below the transformation range, was adopted. One type of the samples used was a standard round test bar ½ inch in diameter. This type of sample, however, never could be converted into single crystals: every time they were treated as described, they would become either coarse-grained on the outside and fine-grained at the center, or vice versa. In one case, the sample had a perfect single crystalline shell, about $\frac{1}{16}$ inch thick, filled with fine-grained material. The explanation of this phenomenon was that in stressing this type of samples in a tensile test

machine, the stress distribution along the diameter of the sample was not uniform. On the other hand, as pointed out by Mr. Kenyon and other authorities, to promote abnormal grain growth a very definite amount of plastic deformation, followed by subjecting to a very definite temperature, is necessary. The actual values of these two factors depend upon each other. Apparently, in our samples, coarse-grained at the surface, the conditions favorable for the abnormal grain growth existed only there, and, in the samples, coarse-grained at the center, conditions were reversed.

This is another illustration of Mr. Kenyon's statement that "there is a critical range of cold working which will result in abnormal grain growth upon annealing".

Author's Closure

Mr. Brophy has raised the question of the effect of segregation on grain-size. The micrographs which he has submitted are understood to be at a magnification of 250 diameters. In our judgment the structures shown in his Figs. 1, 2, and 4 are unsuitable for deep drawing steel sheets. The grains in Fig. 1 are considerably elongated and indicate under-annealing while the coarseness of those in Fig. 2 would very likely cause a roughening of the surface of



a part drawn from such material. The grains in Fig. 4 are also somewhat large.

It is true that there is some difference between the recrystallization behavior of the "skin" and "core" portions of an ingot, but this effect diminishes as the grain-size decreases. The structure shown in Fig. A and B (of this discussion) are typical of those found in the usual run of our deep drawing material. These micrographs were not specially selected. I think they are representative of a large proportion of our low carbon steel sheets for deep

drawing purposes. In both cases the section shows the entire thickness of the sheet from surface to surface. It will be seen that the grains at the surface are slightly larger than at the mid-thickness, but the difference is less on the finer grained specimen. The differences shown by these two samples are not great enough to have any serious effect on the drawing properties.

The discussion by Dr. Ziegler gives a very interesting example of what happens when strain gradients are present in samples of low carbon material before annealing. As he has pointed out, there must have been a difference in the amount of strain across the cross section of his tensile samples, or he would not have gotten such wide variations in grain-size around the surface and in the mid-thickness. His experience has been duplicated by others who have attempted to grow grains of a certain size as an experimental method for obtaining material upon which to conduct such experiments as are necessary to establish the relationship between grain-size and physical properties. The data given in Fig. 8 of the paper were obtained on material that was cold rolled various amounts and then annealed. Probably that method of deformation resulted in a little more uniform strain over the cross section than straining in tension because the grain-size was quite uniform in each sample.

THE INFLUENCE OF GRAIN-SIZE ON MAGNETIC PROPERTIES

By W. E. Ruder

Abstract

The difference of opinion now existing upon the effect of grain-size on magnetic hysteresis loss is apparently due to a number of factors which have not always been considered important by those who have studied the problem. From data presented herein the author concludes that while grain-size does affect the hysteresis loss, other factors, such as purity, method of preparation, size of specimen, and heat treatment, are much more important, and variations in these may easily mask any effect of grain-size alone. He also shows that the hysteresis in single crystals of the same chemical composition and heat treatment may vary as much as 2:1 depending upon the orientation; an effect very similar to that observed a number of years ago with respect to magnetization.

IN reviewing the mass of data bearing upon this general subject, which have accumulated over a long period of years in this laboratory, the first impression is one of confusion. A glance through the literature on the subject, which has increased considerably during the last few years, adds to rather than clears up this confusion. Daeves (4),[1] Von Auwers (5), Honda (6), Von Moos (7), Oertel, Wimmer (8), Yensen (3, 9), and a number of others have contributed valuable data, and yet a general correlation of such data appears to be impossible. Still we find a general agreement that grain-size is important.

In searching for a possible reason for the variation in magnetic quality reported by different investigators, we find differences first, in purity, second, in method of preparation (mechanical), third, in size of specimen, and, fourth, in heat treatment. While the effect of the first is usually recognized, the second and third are given very

[1]The figures in parentheses refer to the bibliography appended.

A paper presented as part of the Grain-Size Symposium of the Sixteenth Annual Convention of the Society held in New York City the week of October 1, 1934. The author, W. E. Ruder, a member of the Society, is associated with the Research Laboratory of the General Electric Co., Schenectady, N. Y. Manuscript received July 5, 1934.

little attention. Von Auwers (5) used a 0.95 millimeter diameter wire, Yensen (9) tested rings 19 by 25 by 25 millimeter deep, while Daeves (4), Oertel (7), Wimmer (8), and others made their experiments on .35-millimeter sheets. It seems to have been generally assumed that the grain-size effect must be due entirely to boundaries and the discontinuities usually associated with them, and so have no relation to the dimension of the grain which is perpendicular to the flux; namely, the cross section of the specimen. For grains considerably smaller than the thickness of the specimen, this effect would be negligible, but for large-grained pure iron or silicon-iron, it becomes very important.

Four rings, all cut from the same ingot of 5 per cent silicon-iron alloy made from electrolytic iron and pure silicon, vacuum fused, and all but #1, 50 millimeter outside diameter by 40 millimeter inside diameter with different thicknesses as shown, were annealed in hydrogen for 24 hours at 1350 degrees Cent. (2460 degrees Fahr.) to remove strains as far as possible and reduce incidental impurities. Ring #1 was cut from the ingot without any working and was 26 millimeter outside diameter, 22 millimeter inside diameter, and 5 millimeter deep.

A summary of the magnetic test data is given in Table I.

Table I

Magnetic Properties of 5 Per Cent Silicon Iron at 10 Kilogausses Compared to Grain Volume and Area

No.	Thickness	H_c	Hyst. Ergs Per cm^3	Residual	Mu Max.	mm^3 Grain Vol.	mm^2 Grain Area
1	5.0 mm	0.115	252	600	16910	525–150–75*	?
2	2.0 mm	0.097	258	1350	16600	29.4	14.7
3	1.0 mm	0.090	300	3100	18600	4.3	4.3
4	0.38 mm	0.186	720	2800	7770	0.238	.70

*Three grains total.

H_c = the coercive force (one-half the width of the hysteresis loop) expressed in oersteds.

μ max = maximum permeability or maximum ratio of induction to magnetizing force.

While #1 and #2 are different sized rings, their section (2 x 5 millimeter) is the same and their hysteresis loss very nearly so. A similar test made on unalloyed electrolytic iron treated in the same way as the iron-silicon rings gave results as shown in Table II.

In both cases the hysteresis increases very rapidly as the thickness of the test sample decreases. To be sure, the surface area of the grain is also a factor, but, as in the case of #3 and #4 electrolytic iron, its effect may be reversed by the thinner specimen.

Table II
Magnetic Properties of Electrolytic Iron at 10 Kilogausses Compared to Grain Volume and Area

No.	Thickness	H_c	Hyst. Ergs Per cm^3	Residual	Mu Max.	mm^3* Grain Vol.	mm^2 Grain Area
1	5.0 mm	0.159	513	7400	16090	1.28(?)	0.64
2	2.0 mm	0.132	442	8100	24380	5.26	2.68
3	1.0 mm	0.171	565	5200	12890	1.50	1.50
4	0.38 mm	0.232	711	4100	9110	0.875	3.1

*Grain count in the case of the electrolytic iron was very unreliable because of the very great irregularity in size and shape of the grains.

In order to check the effect of thickness alone without change in grain-size, rings #2 and #3 of Table I Silicon Iron were cut down from 2.0 to 1.0 millimeter and from 1.0 to 0.5 millimeter respectively, reannealed 12 hours at 1300 degrees Cent. (2370 degrees Fahr.) in hydrogen, and retested.

Table III
Magnetic Properties of Specimens 2 and 3 of Tables I and II

					B = 10000					
	Thickness		Hysteresis		H_c		Residual		Mu Max.	
No.	Before	After	Before	After	Before	After	Before	After	Before	After
2	2 mm	1.0 mm	258	296	0.097	0.100	1350	1450	16600	15300
3	1 mm	0.5 mm	300	348	0.09	0.119	3100	2650	18600	14500

A careful check of grain-size before and after the cutting and reannealing revealed no change in surface grain-size or shape. Reduction in thickness alone increased the hysteresis to almost the same value of the #3 ring in the first test and increased the #3 ring some 16 per cent.

Let us now consider the possible effect of mechanical treatments. Two Epstein samples of a commercial grade of silicon steel, not from the same heat, but having ladle analyses very nearly the same:

A. Hot-rolled Si-3.10, Mn-0.10, S-0.025, P-0.008, C-0.04
B. Cold-rolled Si-2.96, Mn-0.09, S-0.022, P-0.01, C-0.05

were annealed in hydrogen at 1100 degrees Cent. (2010 degrees Fahr.) for 8 hours.

Table IV

	Watt Loss B = 10000			
	Total	Hyst.	Eddy	Grain Area mm^2
A	0.497	0.299	0.198	1.14
B	0.456	0.264	0.192	0.064

These are but two samples selected from a large number of similar tests showing that grain-size, as such, bears an exactly opposite relation to hysteresis loss. Sample A, with a grain area about 18 times that of Sample B, has a hysteresis loss 11.7 per cent higher. Heat treatment is also an important consideration, as may be seen from the following examples:

Three Epstein samples of 0.014 inch transformer sheet, 4.33 per cent silicon, box annealed for 4 hours at 800 degrees Cent. (1470 degrees Fahr.) and slowly cooled, pickled, and tested at B = 10000 60 cycles:

Table V

	Watt Loss B = 10000			
	Total	Hyst.	Eddy	Approx. Grain Area
1	0.447	0.329	0.118	4.0 sq.mm.
2	0.512	0.392	0.120	1.0 sq.mm.
3	0.521	0.424	0.097	0.25 sq.mm.

These show an increased hysteresis and a decreasing eddy loss with diminishing grain-size. The samples were then reannealed for 24 hours at 1200 degrees Cent. (2190 degrees Fahr.) in hydrogen to remove all strains and reduce impurities. After this treatment tests were:

Table VI

	Watt Loss B = 10000			
	Total	Hyst.	Eddy	Approx. Grain Area
1	0.477	0.343	0.134	1½ sq.mm.
2	0.487	0.353	0.134	1½ sq.mm.
3	0.416	0.272	0.144	1½ sq.mm.

Complete recrystallization had taken place so that the average size of #2 and #3 had increased while that of #1 had decreased. Rearranged into three new samples according to grain-size after recrystallization:

Table VII

Watt Loss B = 10000			
Total	Hysteresis	Eddy	Approx. Grain Area
0.474	0.358	0.116	2.0 sq.mm.
0.454	0.329	0.125	0.60 sq.mm.
0.437	0.316	0.121	0.45 sq.mm.

While grain area given here is probably far from accurate, it represents the best possible selection of etched strips with measurements made on a few of representative size. Here we again have a reversal of relationship, the larger grain having both higher total loss and higher hysteresis loss.

A series of seven rings 2¾ inch inside diameter by 3¾ inch outside diameter were cut from a commercial lot of motor grade 0.014 inch sheet, 2½ per cent silicon, annealed for 2 hours at 850 degrees Cent. (1560 degrees Fahr.), pickled, and assembled into samples of thirty laminations each. Magnetic test results at B = 10000 follow:

Table VIII

No.	H_c	Residual	Hyst. Ergs/cm³	Approx. mm² Grain Area
1	0.85	8700	2720	.04
2	0.63	8100	2090	.06
3	0.62	7925	2030	.08
4	0.47	6500	1520	3.0
5	0.38	5400	1310	1.0
6	0.48	5500	1510	4.0
7	0.40	4400	1365	4.0

Here, with one or two exceptions, we find a fairly good correlation; that is, coercive force, residual, and hysteresis all decrease with increasing grain-size. No. 5 is an exception, and examination of its structure showed the grains to be smaller but much more regular in shape than those of #4, #6, or #7. These rings were then reannealed in hydrogen for 24 hours at 1200 degrees Cent. (2190 degrees Fahr.) and retested.

Table IX

No.	H_c	Residual	Hyst. Ergs/cm³	Approx. mm² Grain Area
1	.198	2750	685	mixed .5–6.3
2	.184	3300	590	7.2
3	.192	2500	690	10.0
4	.214	2550	751	3.8
5	.212	2950	703	5.6
6	.190	2500	659	12.5
7	.216	2550	725	25.0

After this heat treatment the grains were all large and the various laminations in each sample were not all the same; hence we could hardly expect any accurate reflection of grain-size. The lowest hysteresis ring, #2, could readily be picked out, after etching, because of its greater brilliancy, possibly due to preferred orientation

with a 1 0 0 plane parallel to the surface. The important point brought out by these examples of the effect of annealing is that the apparent effect of grain-size after the first anneal in each case was probably caused more by imperfect annealing of parts of the sheet than by any real grain-size effect, for a reduction of hysteresis of from 2 to 4, to 1 by reannealing is greater than can be accounted for by grain-sizes of this order of magnitude.

While a number of workers have written about the effect of orientation on permeability, little is to be found about its effect upon hysteresis loss. Honda and Kaya (6) and Gerlach (13) mention the fact that single crystals have very low hysteresis loss. Single crystals of silicon-iron alloy used in our previous experiments (10) on permeability effects were also measured for hysteresis loss; first in the condition as tested at that time (annealed in vacuum 1000 degrees Cent., and then after reannealing for 24 hours at 1400 degrees Cent. (2550 degrees Fahr.) in hydrogen. A re-examination of the structure after this reanneal showed no visible change. The strips were 0.021 inch thick by 0.5 inch wide and 10 inches long.

Table X

No.	H_c 1st Anneal	H_c 2nd Anneal	Ergs/cm³ Hysteresis 1st Anneal	Ergs/cm³ Hysteresis 2nd Anneal	H@B=10000 1st Anneal	H@B=10000 2nd Anneal	% Decrease in W_h by Reann.
1	0.24	0.15	836	488	44	48	41.8
2	0.290	0.17	920	623	83	83	32.2
3	0.245	0.16	759	548	68	66	27.8
3-1	0.270	0.18	826	638	72	71.5	21.7
4	0.375	0.20	1155	661	76	71	42.7
5	0.200	0.12	592	365	11	6.25	38.4
5-1	0.230	0.14	712	379	7	5	46.7
6	0.240	0.14	772	440	69	65	43
7	0.170	0.14	545	391	20	16	28.3
9	0.200	0.13	687	416	9	7	39.4
10	0.150	0.11	554	317	3	2.14	42.8
10-1	0.180	0.105	504	324	3.09	1.35	35.7

Strips #3 and #3-1, #5 and #5-1, and #10 and #10-1 were cut in pairs from the same original single crystals which were each wide enough to make two of the 0.5-inch strips.

Single crystals, then, also vary in hysteresis as much as 3.6 to 1 under different conditions of anneal and orientation and as much as 2 to 1 with the same heat treatment and different orientation. Moreever, there appears to be a relationship between hysteresis loss and orientation similar to that existing between permeability and orientation.

The crystal orientation was determined by X-ray analysis on 5 of the 12 samples (Nos. 1, 2, 4, 6, 10); the angles found are correct within 2 degrees. For the rest the orientation was found by an optical method with an accuracy of about 5 degrees. The specimens were first etched in nitric acid (1 part) plus water (10 parts). Every strip was then mounted on a wooden support and the direction of maximum light reflection marked by a pin. Three directions of maximum light reflection were found in all except one case, these directions forming approximately right angles with each other. They are perpendicular to the three cube faces, and their position can be defined by the angles they form with a system of coordinates fixed with respect to the strip. The x-axis lies parallel to the long axis of the strip, the y- and z-axes in the conventional direction. The orientation is thus defined by the angles which the normals to the faces of an elementary cube made with the coordinate axes, α being the angle between the normal and the x-axis.

Table XI
Angle Between Crystal Axes and Coordinate System

Crystal	100 Axis			010 Axis			001 Axis		
No.	Qu.	α_1	β_1	Qu.	α_2	β_2	Qu.	α_3	β_3
1	3	85	74	1	*31*	62	4	60	33
2	3	79	23	4	49	68	1	*42*	84
3	1	80	58	2	*40*	56	3	58	61
3-1	3	60	65	1	*72*	60	2	*43*	56
4	2	76	29	4	*35*	87	1	60	58
5	2	75	58	3	*15*	83	4	89	40
5-1	2	86	42	4	*77*	64	1	*18*	73
6	3	82	86	2	46	45	1	*45*	45
7	1	86	68	3	*26*	71	4	71	30
9	2/3	90	65	4	66	36	1	*28*	65
10	2	80	57	4	*11*	80	3	84	37
10-1	3	80	60	2	*10*	86	1	86	35

The smallest value of α for every strip is italicized. In the text the symbol λ_1 is used for the cosine of this value.

In Table XI the angles between the face perpendiculars (which represent at the same time the direction of the (1 0 0) axes and the x- (angle α) and y-axes (angle β) are given. Each angle is the average of three separate determinations of maximum reflection. The figures in the column "Qu" indicate in which quadrant the particular crystal axis is located. Table XII shows the coercive force and the losses for the different strips.

Hysteresis, according to the present view as advanced by Heisenberg (16), and Becker, is caused by nonhomogeneous strains in the material. A perfect single crystal should, therefore, show no

Table XII
Hysteresis Loss and Coercive Force Losses for Different Strips

Crystal No.	H_e Oersteds		W_h Watts/lb/cy	
	Before	After	Before	After
1	0.24	0.15	0.00500	0.00290
2	0.290	0.17	0.00547	0.00371
3	0.245	0.16	0.00453	0.00327
3-1	0.270	0.18	0.00494	0.00380
4	0.375	0.20	0.00690	0.00395
5	0.200	0.12	0.00353	0.00217
5-1	0.230	0.14	0.00425	0.00226
6	0.240	0.14	0.00460	0.00262
7	0.170	0.14	0.00325	0.00233
9	0.200	0.13	0.00410	0.00248
10	0.150	0.11	0.00330	0.00189
10-1	0.180	0.105	0.00300	0.00193

hysteresis. These samples, however, still show hysteresis. We might conclude that there still are strains left in spite of the thorough anneal. In addition to that, the hysteresis varies through a range of 1:2 in the different samples, although the crystals have been treated all in the same manner. A relation to the crystal orientation as the only other known variable could be suspected (10).

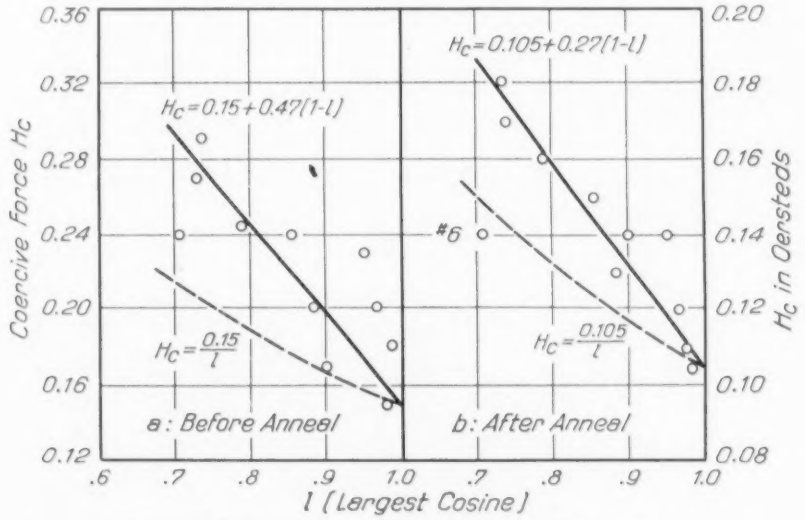


Fig. 1—Coercive Force Before and After Annealing.

In experiments on twisted wires, Tonks and Sixtus (15) showed that in a combined longitudinal and circular field where the angle between the applied field and the induction could be changed at will, it was observed that the coercive force was proportional to the cosine of the angle between field and induction. In other words, only the component of the field in the direction of the induction vector is

effective in causing reversal of induction. In these single crystals Dr. Sixtus suggests that conditions are fundamentally similar, but more complicated than in twisted wires because instead of one there are, in general, three directions along which components of the induction will lie. These directions of easy magnetization are parallel to the three edges of an elementary cube of the crystal and the components of induction lie in these directions up to values of resultant induction much above 10,000 gauss, at which the hysteresis loss was determined. Now if there were only one direction of easy magnetization in these silicon-iron single crystals, we should probably obtain just such a dependence of H_c and W_h on the cosine of angle between I and H. With three directions of easy magnetization, one might assume that the direction in which the greatest part of the total induction lies will determine H_c. This direction is that of the axis which makes the smallest angle with the magnetic field (or the axis with the largest direction cosine), say, l_1.

If H_c or W_h for the annealed strips, however, is plotted against l_1 (Fig. 1), most points, except Nos. 4 and 6, lie close to a straight line which can be expressed by the equation

$$H_c = H_o + A(1 - l_1) \qquad (1)$$

with the constants (after reanneal) $H_o = 0.105$ oerst.
$A = 0.27$ oerst.

The reason for the increasing discrepancy between theoretical curve (H_o/l_1) and experimental values may lie in the observation that in a twisted wire and in a crystal with only one preferred orientation, the internal field structure is homogeneous since the I-vector has the same direction in all the elementary districts. In these single crystal strips, however, the districts where I has the a_2 or a_3 direction act as barriers for the propagation of magnetization in the a_1 direction. An additional field has to be applied in order to overcome those disturbing fields, and it has to be the greater the greater the total volume of the a_2 and a_3 districts, i. e., the greater a_2 and a_3 are or the smaller l_1 is.

A possible reason for the failure of strip #6 to conform with Eq. (1) lies in the fact that it shows a distinguished crystal orientation with one axis almost perpendicular to the strip surface and the other two at an angle of 45 degrees to the field. It may well be that under these simple conditions propagation will proceed with a smaller field than under ordinary conditions. No reason for the

discrepancy in the case of strip #4 can be suggested, except that since its value is above the curve, some strain or impurity effect may exist in it.

In Fig. 1, H_c before reanneal is also plotted; in Fig. 2 the hysteresis loss before and after reanneal.

This relationship also offers an explanation for the often observed dependence of hysteresis loss on angle between field and

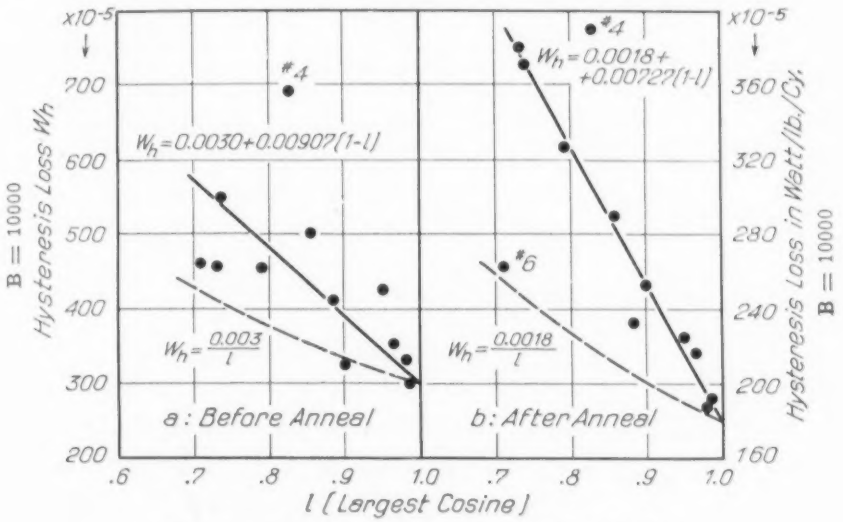


Fig. 2—Hysteresis Loss Before and After Annealing.

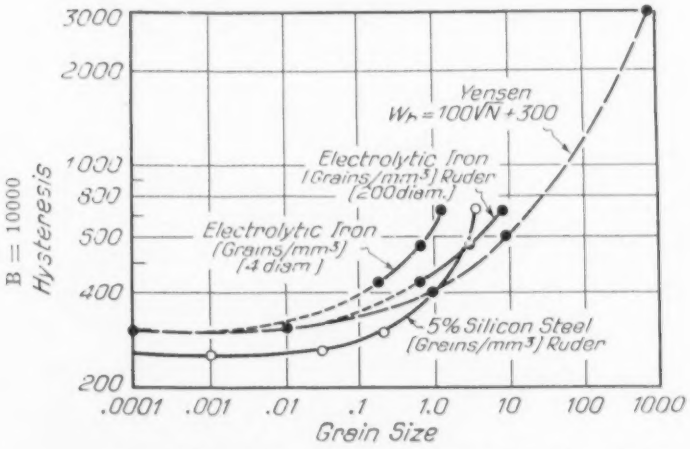


Fig. 3—Hysteresis—Grain-Size Curves on Electrolytic Iron and 5 Per Cent Silicon Steels.

rolling direction in hot and cold-rolled sheets. The working results in a more or less uniform preferred orientation which is quite marked even after annealing. In hot-rolled sheets we have found a preferred

crystal orientation with a (1 1 0) axis parallel to the rolling direction and a (1 0 0) plane in the surface, and it may be expected from the foregoing results that the hysteresis should be higher parallel to the direction of rolling than under 45 degrees to this direction. This we have found to be true only under certain rolling conditions.

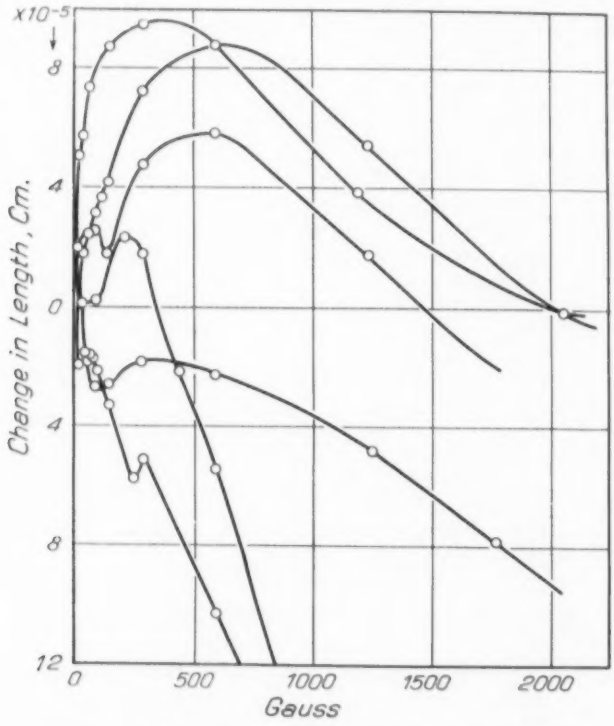


Fig. 4—Variation of Magnetostriction with Orientation.

Cold-rolled sheets present a different type of preferred orientation depending upon the degree of cold reduction and also show minimum hysteresis along the direction of preferred orientation.

Is there then no relationship between grain-size and hysteresis loss, other things being equal? Yensen (9), after a very careful inquiry, answers in the affirmative. His rings of electrolytic iron were all of comparatively large section (3 by 5 millimeter). Those given in Table II of this paper were of varying cross section, but for equal grain-size and equivalent thickness give about the same hysteresis value. Since practically all of our grains went through the thickness of the sheet (except #1 electrolytic iron), the volume in cubic millimeters was readily determined, and Fig. 3 shows a fairly good agreement with Yensen's curve, both for electrolytic iron and

silicon steel if *volume* of grain rather than *area* of grain is used. It will also be observed that even the lowest hysteresis value (317) in the thin (0.53 millimeter) single crystal strips is not so low as the lowest value (252) for the thicker rings, but is nearly the same as that of the ring of nearest thickness.

If then the volume of the grain must be taken into account in its relation to hysteresis loss; some reason must be found to account for the grain-size effect other than or in addition to that of boundary effects, such as impurities, grain-fragmentation or incomplete crystal formation. Certainly boundaries would have no effect in single crystals. While the effect of orientation might be expected to average out in the core of polycrystalline samples, the variation of magnetostriction with orientation is still a possible factor (Fig. 4), (11) (14), although at B = 10000 magnetostriction is very small. Variation of hysteresis loss at higher densities where change in length and volume with flux is more pronounced, shows some indication that this factor is important, but data available are yet too meager to draw any conclusions.

Hysteresis, then, does depend upon grain-size, but to a much lesser extent than upon a number of other factors, such as purity, heat treatment, mechanical treatment, and shape of sample tested. Hysteresis also bears a distinct relation to orientation, which in turn has an important bearing upon the rolling direction effects observed in polycrystalline sheets and strips.

Acknowledgment is made of the assistance of Dr. K. Sixtus, who determined the orientations of the single crystal strips and worked out the equation for their relation to hysteresis loss, and to B. M. Smith who made the magnetic tests.

References

1. W. E. Ruder, *Transactions,* American Institute of Mining Engineers, Vol. 47, 1913, p. 569.
2. Mr. Thompson, *Proceedings,* American Philosophical Society, Vol. 31, 1916, p. 357.
3. T. Y. Yensen, *Transactions,* American Institute of Electrical Engineers, Vol. 43, 1924, p. 145.
4. Mr. Daeves, *Stahl und Eisen,* Vol. 44, 1924, p. 1283.
5. Mr. Von Auwers, *Zeitschrift für technische Physik,* Vol. 6, 1925, p. 578.
6. Kotaro Honda and Seiji Kaya, "On the Magnetisation of Single Crystals or Iron," Science Reports of Tohoku Imperial University, Series 1, Vol. XV, No. 6, 1926, p. 721.
7. Messrs. Von Moos, Oertel and Scherer, *Stahl und Eisen,* Vol. 48, 1928, p. 477.

8. Messrs. Wimmer and Werthebach, *Stahl und Eisen*, Vol. 54, 1934, p. 385.
9. T. Y. Yensen, *Metals and Alloys*, Vol. 1, 1930, p. 493.
10. W. E. Ruder, "Magnetization and Crystal Orientation," TRANSACTIONS, American Society for Steel Treating, Vol. VIII, 1925, p. 23.
11. Same as Reference (6), p. 755.
12. Mr. Webster, *Proceedings*, the Royal Society, Vol. 107, 1925, p. 496.
13. Mr. Gerlach, *Physikalische Zeitschrift*, Vol. 26, 1925, p. 914.
14. Mr. Williams, *Testing*, Vol. I, 1924, p. 190.
15. Messrs. Tonks and Sixtus, *Physical Review*, American Physical Society, Vol. 43, 1933, p. 76.
16. Mr. Heisenberg, *Zeitschrift für Physik*, Vol. 49, 1928, p. 619.

DISCUSSION

Written Discussion: By T. D. Yensen, Manager, Magnetic Division, Research Laboratories, Westinghouse Electric and Manufacturing Company, East Pittsburgh.

The conclusion at which I have arrived on the basis of experimental evidence has been stated in a number of papers and discussions on the subject, some of which Dr. Ruder has referred to and correctly interpreted. Briefly stated it is this: In bulky samples of iron and iron alloys in which the dimensions of the grains are randomly oriented and approximately equal in all directions, the hysteresis loss (as well as coercive force and maximum permeability) depends upon the grain-size in such a way that the larger the grains the lower the loss. Ruder first called attention to this in 1916, and I supplied quantitative values in 1924. In addition to grain-size, the hysteresis loss depends upon the impurities in the material, chiefly C, O_2, S and P as shown in papers by myself and jointly with Mr. Ziegler. In order to obtain a correct idea of the effect of grain-size, it is essential to eliminate the effect of the impurities, or to keep these very low. Otherwise the effect of the impurities may entirely mask the effect of grain-size. This point has, I think, been one of the main causes of the controversy in regard to this subject during the past 10 years, and it is interesting to note that in a recent paper by Dr. Eilender and his associates in Aachen, I was accused of the error of *not* having taken the impurities into account in trying to determine the effect of grain-size. This, as most of you know, is not the case. I have pointed out repeatedly during the past 10 years that the effect of impurities *must* be eliminated before we can draw conclusions in regard to the effect of grain-size. Differences in language probably account for this misunderstanding.

Now when we come to thin sheets or strips, it has been apparent for some time that the case is not so simple as in bulky samples, that some other factor besides size enters the picture. The high hysteresis loss obtained in ring samples made from thin sheets or in strip samples from ordinary sheet material could not always be accounted for on the basis of impurities and grain-size alone. As soon as we were able to produce a strip or sheet material with grains predominantly oriented with the cube edge parallel to the magnetizing force, we obtained a hysteresis loss comparable to that obtained in bulky samples of similar composition.

Dr. Ruder has done much to help clear up this situation by presenting in his present paper quantitative results in regard to the effect of orientation on the hysteresis loss. I am, however, disappointed in his statement in the next to the last paragraph of this paper in which he states "Hysteresis, then, does depend upon grain-size, but to a much lesser extent than upon a number of other factors, such as purity, heat treatment, mechanical treatment and shape of sample tested." While this statement is undoubtedly true, would it not be much better to confine our statement to one fundamental cause, namely, lattice distortion, because grain-size, which means grain boundaries, impurities, mechanical and thermal treatment all seem to have one fundamental effect in common, namely, to distort the crystal lattice. As we are beginning to think of magnetic properties in terms of the lattice structure, it would seem logical, therefore, to relate these properties whenever possible to the lattice structure.

Written Discussion: By N. P. Goss, physicist in charge of metals research, Cold Metal Process Company, Youngstown, Ohio.

Dr. Ruder's paper should be of great interest to all who are interested in the magnetic properties of the ferro-magnetic materials and should stimulate further research on the various factors (which influence the magnetic properties) that is,

1. Purity
2. Manner in which specimen was prepared (mechanical)
3. Size of specimens
4. Heat treatment

He points out that while one and four are usually recognized, two and three have been given very little consideration. I should like to add that the magnetic properties of ferro-magnetic alloys depend upon the manner in which the hot rolling, cold rolling and heat treating are distributed during processing.

My results show that the watt losses can be reduced to a very small value when material is processed by proper distribution of hot rolling, cold rolling and heat treatment. Another factor of great importance and usually not considered at all is the perfection of crystallinity.

At the present time most authorities believe that the hysteresis in large grain silicon iron specimen is smaller than the corresponding fine grain specimens. Dr. Ruder, after making an exhaustive survey on various specimens, comes to the conclusion that the hysteresis loss does depend on the grain-size (microscopic) to some degree, but that the factors already mentioned are of greater importance.

From the results of my own work I have found that the hysteresis loss becomes smaller as the grain-size decreases provided the final product has single crystal characteristics, therefore, these results are again in agreement with Dr. Ruder's.

In his experiments the test specimens were all annealed at extremely high temperatures in hydrogen for long periods of time. Under such conditions the purity of these iron-silicon alloys were improved and the hysteresis greatly reduced.

I should like to ask whether or not the results would have been the same had the samples been heat treated in a neutral atmosphere, and also what effects would have resulted had lower annealing temperatures been used?

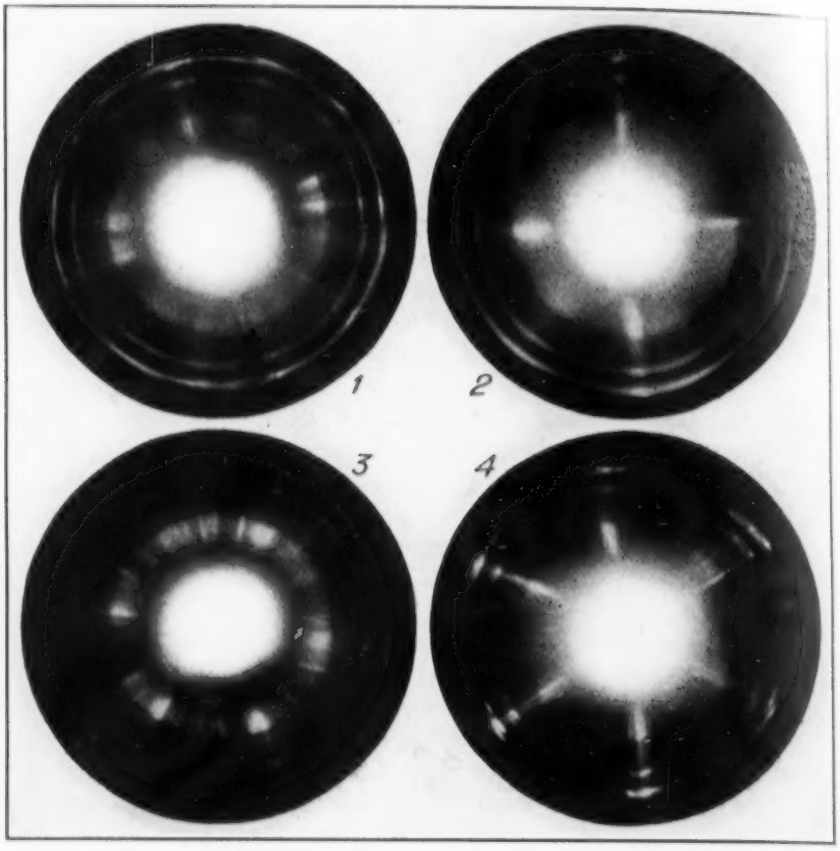


Fig. 1—Hot-rolled Silicon Steel Strip Given an Area Reduction of 62½ Per Cent X-ray Beam Transmitted Normal to Strips Surface.
Fig. 2—Hot-rolled Silicon Steel Strip After Being Given an Area Reduction of 92 Per Cent. The X-ray Beam Was Transmitted at Right Angles to the Surface of the Strip.
Fig. 3—The X-ray Structure of a Hot-rolled Sheet 0.014 Inch After Pack Rolling. The X-ray Was Transmitted Perpendicular to Surface.
Fig. 4—X-ray Structure of a Hot-rolled 0.105 Inch Thick Strip X-rayed as Shown in Fig. 4A.

It is remarkable to find a hysteresis loss of only 680 ergs/cm³/sec after heating an ordinary motor grade of 2½ per cent silicon steel in hydrogen at 2190 degrees Fahr. for twenty-four hours. See Tables VIII and IX and refer ot specimen No. 1. I believe these hysteresis losses are lower than ever reported on this grade of material. I would like to ask whether the results of these experiments would mean that high grade silicon-iron alloys used in electrical machinery will be of relatively low silicon content in the future?

In view of the great reduction in hysteresis it would be of interest to know the value of μ_{max} and $\mu_{10,000-B}$.

In Table IV the possible effect of different mechanical treatments are considered; that is, the cold rolling and hot rolling are compared. However, before the results given can be accepted a complete log of the various steps from the ingot to the finished specimens must be given. The equivalent cold work[1] in the hot-rolled specimens prior to the final heat treatment should be

[1] "Effects of Cold Working on Physical Properties of Metals by R. L. Templin—A. I. M. E., September, 1929.

determined so that a direct comparison can be made to the cold-rolled structure prior to the final heat treatment. It was of interest to know that under certain conditions of cold working the hysteresis in the direction of rolling can be made greater than at an angle of 45 degrees to this direction. I would like to know what the rolling conditions are in order to obtain these results?

On page 1130 Dr. Ruder states that cold-rolled sheets present a different type of preferred orientation depending on the degree of cold working.

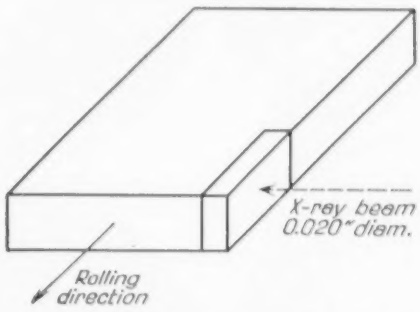


Fig. 4A—Illustrating What is Meant by Principal Direction.

From my own work on the cold working of hot-rolled strip the X-ray shows that only one type of orientation can exist, but becomes of great intensity as the amount of cold work is increased.

I also find that as a rule only a weak fibre structure is developed in the plane of rolling for reductions not exceeding 70 per cent.

Fig. 1 is the X-ray Laue pattern of a specimen given an area reduction of 62 per cent while in Fig. 2 the same material is given an area reduction of 92 per cent. In contrast Fig. 3 is the X-ray diffraction pattern of a hot-rolled sheet ready for the final heat treatment.

The most intense fibre structure is never developed in the plane of rolling but in the direction shown in Fig. 4A. This I prefer to call the principal direction.

Fig. 4 is a diagram made of silicon-iron hot-rolled strip in which the X-ray beam was transmitted in the principal direction; whereas, Fig. 5 is the structure after heat treatment.

Now let us see what happens when this heat treated hot-rolled strip is given various area reductions.

Fig. 6 is the X-ray diffraction pattern after it was cold-rolled from 0.105 to 0.096 inch.

Fig. 7 after cold rolling from 0.105 to 0.056 inches.

While in Fig. 8 is a diffraction pattern after rolling from 0.105 to 0.036 inch and shows conclusively that the type of prefered orientation remains the same, but the degree of orientation does vary.

Grain-Size or Crystal Structure

Now it may be argued that the microscope does not measure the true grain-size, and certainly does not tell us anything about the internal structure. The

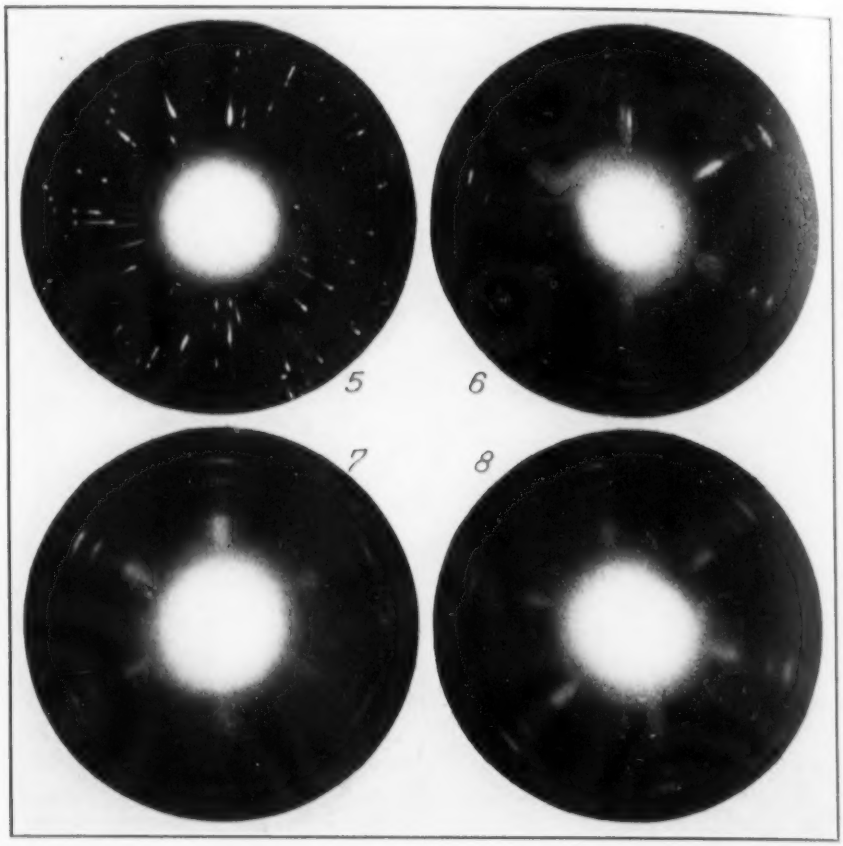


Fig. 5—The X-ray Structure of Hot-rolled Strip After Being Heat Treated and X-rayed in Principal Direction.

Fig. 6—The Heat Treated Hot-rolled Strip Shown in Fig. 5, After Being Cold-rolled from 0.105 to 0.096 Inch.

Fig. 7—Heat Treated Hot-rolled Strip After Being Cold-rolled from 0.105 to 0.056 Inch.

Fig. 8—Heat Treated Hot-rolled Strip Cold-rolled from 0.105 to 0.036 Inch.

grains viewed under the microscope may actually be composed of many secondary grains.

This would mean that the structure is mosaic instead of being composed of real single crystals or grains. F. Zwicky defines a mosaic crystal as a real crystal composed of tiny crystal fragments which are slightly displaced or inclined relative to each other. That such grain distortions can be produced during heat treatment is presented by Professor Nusbaum and myself at this meeting. ("Grain Distortion During Heat Treatment.")

Radial asterism as found in X-ray diagrams tells us about the internal condition of the grains and which in the past has been considered to be an indication of stresses or strains existing in such materials. I will now show a few X-ray diffraction patterns illustrating the difference between real crystals and mosaics.

Fig. 9 is the X-ray Laue diagram of a pure electrolytic iron which was cold-rolled from 0.050 to 0.002 inches and annealed at 800 degrees Cent. This structure represents a nearly perfect crystal structure in that all of the grains

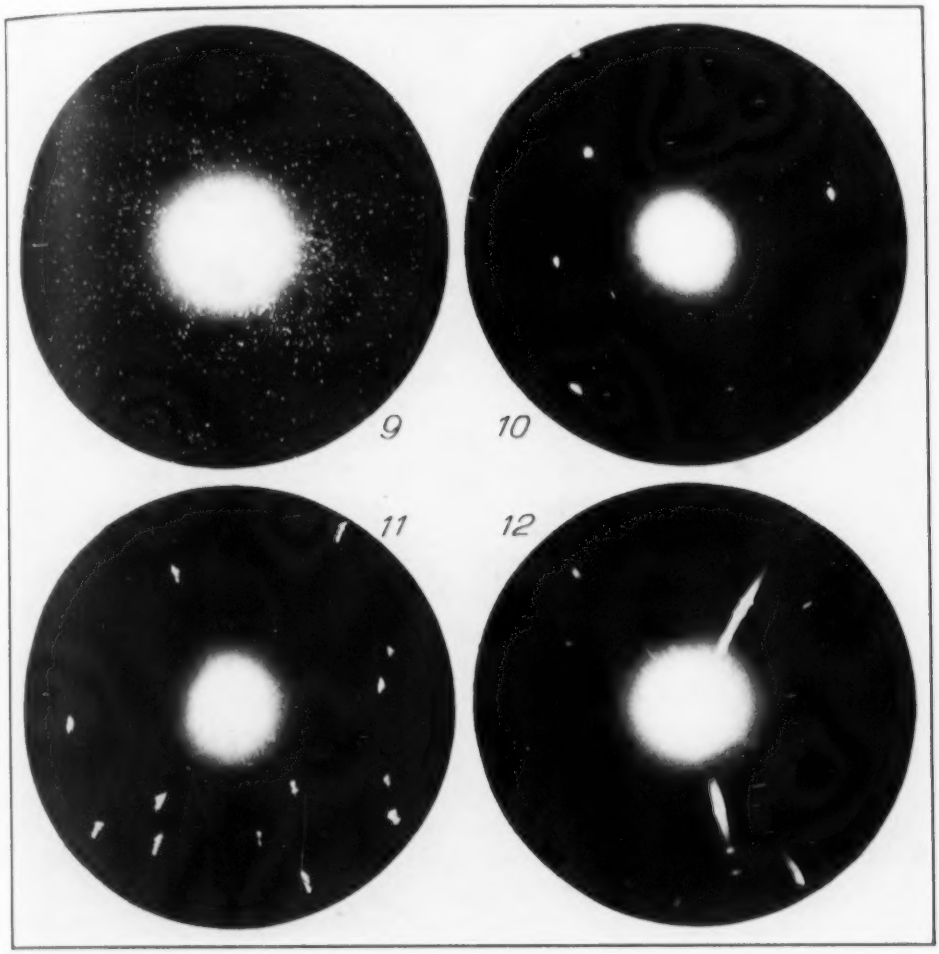


Fig. 9—Pure Iron Strip Cold-rolled from 0.050 to 0.002 Inch and Heat Treated at 800 Degrees Cent.
Fig. 10—A Single Crystal of Silicon Steel.
Fig. 11—One Form of Mosaic Structure.
Fig. 12—A Single Grain From a Small Silicon Steel Ingot, It Was Equiaxed Yet the X-ray Showed It to Have a Fibrous Structure.

are free of grain distortion and the metallic aggregates are composed of small single crystals free of internal imperfections. I have found that as a rule such structures can only be produced in relatively thin materials.

In Fig. 10 the crystal structure of a single crystal of silicon steel grown in the strip only 0.010 inches thick at 2200 degrees Fahr. is shown. It is nearly perfect in crystal structure and represents a single crystal.

Fig. 11 is an X-ray spectrogram of one type of mosaic crystal structure. You will note that each of the diffraction maxima are actually composed of several lines parallel to each other. In other words, what was thought to be a single crystal was actually composed of several having approximately the same orientation.

In Fig. 12 a single grain removed from a small silicon steel ingot is shown and represents another type of mosaic in which the structure has a fibrous

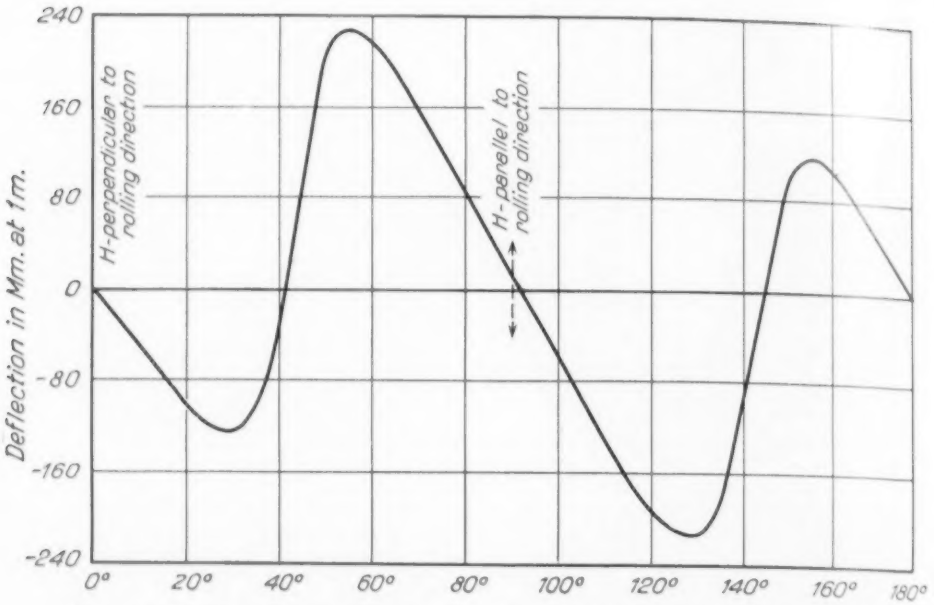


Fig. 13—Torsion Magnetization Curve of a Single Crystal, Showing that Easy Magnetization is Possible Only in Two Directions.

character. This grain appeared to be a single crystal about ¼ inch in diameter, therefore, in view of the X-ray diagrams which I have shown; and before Dr. Ruder can conclude that hysteresis and grain-size are not related he must show by X-ray spectra that all of specimens tested were composed of real grains and not mosaic or crystals having internal imperfections. In other words, the hysteresis loss is related to the internal structure of the grains rather than to the grain-size as observed under the microscope.

Single Crystals

Dr. Ruder's results made on the single crystal strips described on page 1125 and which he had originally presented in TRANSACTIONS in July, 1925, are of the utmost importance, especially in that hysteresis depends on the direction on which the crystal is magnetized.

I will offer an explanation for the variations found in hysteresis and permeability in various crystallographic directions. It can be shown by the use of a Torsion Magnet that single crystals of any ferro-magnetic material are characterized by a magnetic moment, the magnitude of which is given by the moment of the couple expressed by the Vector quantity M x H where M is the magnetic moment and H is the field strength. In other words, when a single crystal disk is suspended in the field of a Torsion Magnet and rotated it will experience a torque, the magnitude of which is the tensor of M x H and depends only on the moment of the elementary magnets given by Σ (m) ds = M where M represents the pole strength and ds is the distance between the elementary dipoles. The magnitude of the magnetic moment is related to the intensity of magnetization for any given field strength H. If we have a volume distribution of magnetic material of volume dv then the resultant

moment of all the elementary magnets in the element are replaced at a single vector I = M divided by dv and is defined as the intensity of magnetization in some given crystallographic direction and is the moment per unit volume of I dv = M. This means that single crystals can only be magnetized with ease in certain crystallographic directions. In the case of higher silicon single crystals it has been shown by Honda, Kunz, Ruder and Yensen that the [100] is the direction of easiest magnetization followed by [110.] Dr. Ruder has conclusively shown that the hysteresis is lowest in the [100] direction.

When such crystals are tested in the Torsion Magnet the greater the deflections the greater the intensity of magnetization for any given field strength H, that is, the greater the tensor value of I dv x H the greater is the permeability and the lower the hysteresis. By using Poissons equation it can be shown that $\triangle (H + 4 \pi I) = 0$ the new Vector $H + 4 \pi I = B$ is the magnetic induction. This means that for any given H, B depends on I which in turn is a function of the magnetic moment per unit volume. In the case of single silicon-iron crystals removed from hot and cold-rolled sheets in which large crystals can be grown torsion magnetization curves of the type shown in No. 14 are always found. These curves are always characterized by 4 maxima one of which are twice as high as the other two. This means that the permeability is twice as great in one direction as in the other and the hysteresis varies in about the same ratio.

The question I would like to ask is, does the equation given on page 1128 take into consideration that single crystals can only be magnetized with ease in two directions in which the paths are at right angles to each others?

Written Discussion: By N. A. Ziegler, Power Use Research Department, The West Penn Electric Company, Pittsburgh.

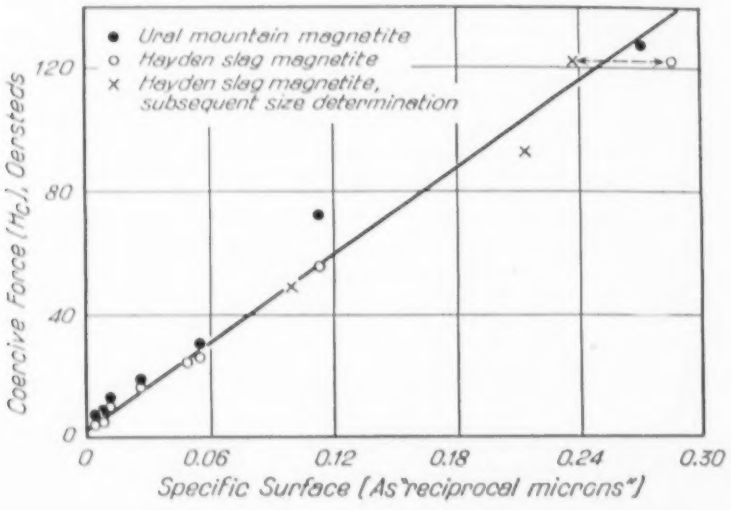
It is gratifying to hear from Dr. Ruder a confirmation of my own opinion that "hysteresis does depend upon grain-size." It is also very interesting to learn that "hysteresis bears a distinct relation to orientation." This is a novel idea which requires a very careful consideration and further investigation.

In regard to the numerical data presented by Dr. Ruder, his own opinion that purity of a sample is, perhaps, the most important factor determining its magnetic properties, should be amplified. In a paper presented at this meeting, jointly with Dr. Yensen, the effect of carbon (all other factors being eliminated) and oxygen (all other factors being eliminated) on hysteresis and maximum permeability of "pure" iron is demonstrated. It is clear that a difference of even a few thousandths of one per cent of these two elements has a profound effect on the magnetic properties of unalloyed iron. Carbon has a similar effect on the magnetic properties of silicon alloys, which may be judged from the earlier results by Dr. Yensen, and from our results which will be presented at a later date.

A variation as small as a few thousandths of one per cent of carbon may easily be introduced to a magnetic sample by the heat treatments, and, for this reason, it is felt that unless a complete chemical analysis of every magnetic sample, after its magnetic testing is completed, is given, its history is incomplete. This is particularly true about carbon which must be determined in every magnetic sample individually with an accuracy of at least $\pm$ 0.001 per cent.

Written Discussion: By R. S. Dean, chief engineer, metallurgical division, U. S. Bureau of Mines, Washington, D. C.

The conclusions from this very careful work by Dr. Ruder do not seem to warrant a definite connection between grain-size and magnetic properties. Our work in the Bureau of Mines on mineral powders has led us to such definite and clear cut relationships of this kind that we are hopeful that such relations do exist for steel but may require a revision of our ideas on grain-



size. A revision which in point of fact is being made by the steel metallurgists generally as indicated by the papers of this symposium.

Dr. Gottschalk has measured the coercive force of very carefully sized magnetite powders, keeping the packing density closely constant. His results are shown in above figure. The linear relation between specific surface and coercive force is quite evident.

In view of the clear cut nature of these results we are inclined to think that coercive force is a definite measure of what may be termed "internal surface" which may result from grain boundaries as ordinarily described, precipitated phases, glide planes, or other discontinuities. The coercive force in steel may then be a measure of grain interface if we define a grain as a continuous "homogeneous volume of metal lattice." It is suggested that such homogeneous volumes are the determining factor in potential grain-size after hardening. Some experiments are under way to test this suggestion.

Author's Reply

The remarks made by Dr. Yensen and Mr. Ziegler, referring to their quantitative results on the effect of purity of the material on hysteresis loss and on permeability, are well taken. To date no one has been able to challenge any of the results which they have obtained through their careful and painstaking research. The only question which I have raised in my present paper concerns the validity of expressing the effect of grain-size on magnetic properties in terms of surface area as Dr. Yensen has done in his previous

publications. I have attempted to show that grain volume rather than grain surface is the only fundamental way in which this relationship can be shown. I regret his disappointment in my general statement about the effect of grain size on hysteresis and the only reason I can give for making it is that it is true. To generalize further and say that lattice distortion is the fundamental cause would be going a little further than the evidence justifies. I am inclined to believe that stresses set up by magnetostriction, which differs with the orientation from grain to grain, may be the fundamental cause for the grain-size effect in reasonably pure metals or alloys. If that is proven to be so, we might generalize grain-size effect as being one of crystal lattice distortion.

Dr. Dean presents interesting data on the effect of grain-size of magnetite on coercive force, and I can only share with him the hope that such a relationship as we find in iron-silicon alloys and pure iron can be correlated with Dr. Gottschalk's results. We have done some work on powders of brittle silicon-iron alloys, but our data are not comparable to those reported by Dr. Dean. Perhaps a study of the metal powders made up in a similar way to the magnetite might prove interesting.

A partial reply to Mr. Goss' remarks has already been presented in my discussion of his paper read earlier in the week. The reason for annealing my samples at a high temperature and in hydrogen was to free them as much as possible from the effect of residual strain and incidental impurities, and the results would not have been the same had such treatment not been used. The low hysteresis losses reported on some of my 2½ per cent silicon sheets do not necessarily mean a trend away from the higher silicon alloys. The choice of the silicon content depends on the variety of engineering factors which must be considered. The permeability values asked for were not unusual. The maximum permeability was about 7000 and the permeability at 16,000B was about 400. The conditions under which 45-degree properties are superior to with-rolling-direction properties occur after hot rolling and not cold rolling.

While Mr. Goss may well argue from his X-ray pictures that no preferred orientation exists in annealed cold-rolled strip, our own results obtained by optical methods show a prevalence of distinct types of preferred orientation varying somewhat with the degree of cold rolling, and these orientations are directly related to the magnetic properties of single crystals similarly oriented. If mosaics exist in the single crystals or poly-crystals we have studied, they will certainly be expected to have an important effect on magnetic properties, as we believe that any discontinuity in the crystal lattice would do this. We have reason to believe, however, that under the conditions of mechanical and heat treatments used our samples were probably quite free from such disturbances.

FACTORS DETERMINING THE IMPACT RESISTANCE OF HARDENED CARBON STEELS

By Howard Scott

Abstract

Recognizing the possibility of greater significance attaching to impact tests on hard steels than on soft, impact tests were made on a variety of fully hardened carbon steels of medium carbon contents, using a Charpy type specimen with a special notch. No pronounced maximum in impact resistance at low tempering temperatures was disclosed by this test. Complete solution of ferrite was found to be essential, however, for full development of impact resistance.

Comparing various compositions as quenched from 830 degrees Cent. and tempered at 300 degrees Cent., nearly all impact values corrected for hardness differences fell within one of two widely separated groups. Laboratory melts deoxidized with manganese and silicon alone fell in the low impact group as did one commercial steel. Certain additions of aluminum and vanadium brought the impact values consistently in the high group.

Aluminum treated steels had a fine austenitic grain size and gave high impact values at hardness values up to 600 Vickers (55 Rc.) at least, while coarse grained steels lost impact resistance above 400 Vickers (41 Rc.).

Observations on highly purified steels indicate that greater need for additions which restrain grain growth will accompany advances in this direction.

Introduction

DESPITE the great amount of work that has been done on the notched-bar impact test, opinion remains divided concerning its significance with regard to service failures. Davidenkoff (1)[1] ably supports the contention that standard tests have no implications as to allowable impact stresses in large structures. Nevertheless Greaves (2) finds from extensive practical experience with large

[1]The figures appearing in parentheses refer to the bibliography appended to this paper.

A paper presented as part of the Grain-Size Symposium of the Sixteenth Annual Convention of the Society held in New York City the week of October 1, 1934. The author, Howard Scott, a member of the Society, is associated with the Research Laboratories of the Westinghouse Electric and Manufacturing Co., East Pittsburgh, Pa. Manuscript received June 30, 1934.

forgings that steels falling below a certain impact value determined by a conventional test are decidedly liable "to fail prematurely under localized stress." Their divergent views, however, are based on tests of soft steels, that is, steels in the pearlitic or sorbitic condition. Tests on hard steels, martensitic or troostitic, may on the other hand lead to widely different conclusions.

Martensitic and troostitic steels are widely used in service where impact loading is the rule, for example, in gearing, springs, cutting tools, twist drills and dies. The impact resistance of such steels is of much lower magnitude than that of the softer steels. On this account alone differences in impact resistance should be more significant in them than in high impact steels, because the impact resistance of the softer steels, even though relatively low, is usually adequate to withstand service impacts. Furthermore they are used in sections whose dimensions are comparable to those of standard notched-bar impact test specimens so one may properly expect to find more sharply defined relations between service performance and impact resistance in the harder class of steels.

Anticipating that some useful findings would develop from a study of impact resistance as influenced by composition and heat treatment of fully hardened and moderately tempered carbon steels, such an investigation was started several years ago. The results obtained are now especially pertinent since a close relation between impact resistance and grain size of hardened steels is becoming apparent. (3)

Grain-size in turn is recognized as a major factor determining the ability of hard steel to withstand the punishment of quenching and grinding. (4) It also has a dominant influence on penetration of hardening. (5) The writer has found a coarse grain-size associated with failure under static bending stresses estimated to be safe from tensile tests. Hence notched-bar impact tests on hardened steels have implications as to their ability to survive heat treatment and grinding, liability to service failure and hardenability. This conclusion suggests that the impact properties of hardened steels are deserving of systematic study.

Materials, Treatments and Tests

When undertaking to determine impact resistance of steels harder than 400 Brinell, cutting of the A. S. T. M. standard notch after heat treatment was not considered practicable or desirable.

Cutting of notches before treatment is certainly undesirable because of the danger of adverse effects from possible oxidation, cracking or residual stress concentration in the notch. As a compromise notches of a more convenient shape were ground in the impact test specimens after heat treatment and finished by lapping to remove transverse scratches. A Charpy type specimen 0.394 inch square and 1.58 inches between supports was used with notch 0.062 inch radius and 0.079 inch deep.

Impact tests were made on an Amsler machine of 120 foot-pounds capacity. Three breaks were made on each specimen which was handled in one piece up to the time of testing. The test had adequate sensitivity for the purpose and the results were quite consistent.

Vickers hardness tests were made on every specimen tested in impact. Since the Rockwell hardness scale is more familiar to most readers, conversion valves are given in Table I. These data are taken from a paper by Sawin and Stachrowski. (6)

A variety of steels was tested but most of them fell within the carbon range of 0.4 to 0.5 per cent. Compositions are given in Table II and include analysis of nonmetallic inclusions extracted by the electrolytic method of Fitterer. (7) Aluminum was determined by the phosphate method so that the reported values include at least part of that present as alumina.

Certain of the commercial steels were made under the supervision of Dr. C. H. Herty, Jr., who furnished samples for test. The Furnace practice used in making these steels is fully described elsewhere. (8) These steels are identified in that paper by heat numbers as follows:

Steel Number	Heat Number
94	49212
103	31058
104	33048
105	44100

The 2900 series of steels was made in a high frequency induction furnace as a part of this investigation. Charges weighing 18 pounds were melted in magnesia-lined graphite crucibles. The melting stock was Armco iron. Additions of carbon, manganese, silicon and aluminum were made in the order listed. The ingots were, unless otherwise noted, cast shortly after the last addition into square, tapered, steel molds 10 inches long and forged to 1⅛ inch round and ½ inch square bars, impact specimens being machined from the latter.

Since it was essential for the announced purpose to obtain a completely martensitic structure as quenched, particular attention was given to the problem of a powerful and reproducible quench. Analysis of this problem (9) suggested that a submerged water spray quench would give as fast and reliable a quench as is practicable.

Table I
Comparison of Vickers and Rockwell "C" Hardness from Data of Sawin and Stachrowski

Rockwell "C" Hardness	Vickers Hardness (50-kilogram load)
65	880
60	740
55	615
50	519
45	451
40	392
35	342
30	300
25	266
20	235

A round tank of the dimensions given in Fig. 1 was used with a tap water supply under about 80 pounds per square inch pressure. Etched sections through 1-inch diameter rounds quenched in this bath showed that it gave unusually deep and uniform penetration of hardening and a high degree of reproducibility.

Specimens were heated for quenching in an electric furnace without atmospheric control. A quenching temperature of 830 degrees Cent. (1525 degrees Fahr.) was used except where otherwise noted. Tempering was done in oil and fused salt baths for a period of 40 minutes at temperature.

Tempering Relations

Two commercial steels, Nos. 94 and 95 of Table II which differed widely in impact resistance, were studied in some detail. Impact and hardness values of the steels are plotted against tempering temperature in Figs. 2 and 3. There is a region of rapid rise in impact resistance of both steels on tempering at temperatures between 300 and 400 degrees Cent. (570-750 degrees Fahr.). Since the hardness does not diminish correspondingly, the relation between impact resistance and hardness, if any, is non-linear. Impact resistance, nevertheless, is closely related to hardness, as will be shown presently.

Luerssen and Greene (10) have observed a prominent maximum

in the torsional impact curve of 1.1 per cent carbon steel at about 180 degrees Cent. (355 degrees Fahr.). Indications of a similar effect in bending impact may be seen in Figs. 2 and 3, that is, the curves tend to flatten out after rising to the 180 degrees Cent. (355 degrees Fahr.) tempering temperature. The absence of a drop at

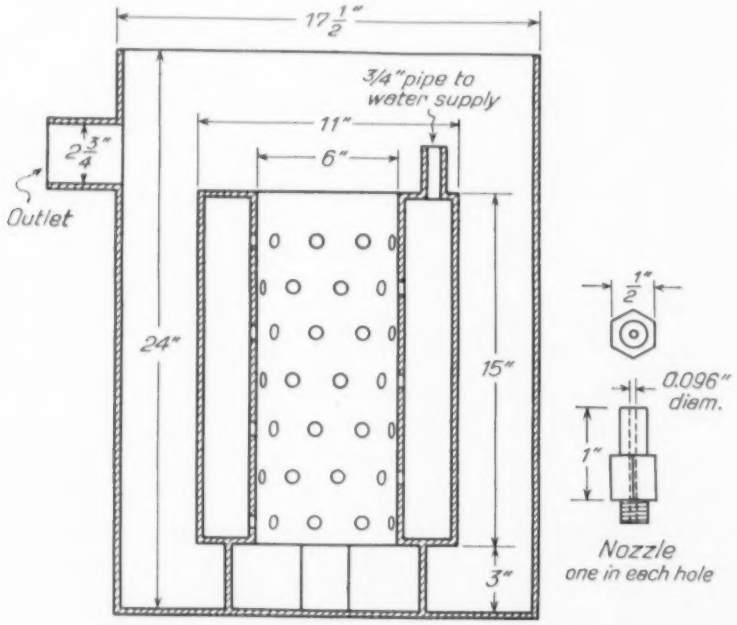


Fig. 1—Submerged Water Spray Quenching Tank Used in Hardening All Test Specimens.

higher temperature, however, is noteworthy in view of the torsion test results. An explanation of this difference between these tests should be sought in the mechanics of the respective types of failures.

For the first step in establishing a broad relation between impact resistance and hardness, these properties for each treatment were plotted against each other as in Fig. 4. This eliminates tempering temperature as a variable considered and permits one to see clearly the effect of quenching temperature. That factor has little influence on the brittle steel, No. 94, while it has a pronounced effect on the tough steel, No. 95. Surprisingly, the low quenching temperatures, 780 and 800 degrees Cent. (1435-1470 degrees Fahr.) were harmful, presumably because of free ferrite. At least free ferrite was not present in No. 94 as quenched from 780 degrees Cent. (1435 degrees Fahr.), but still remained in No. 95 as quenched from 800 degrees Cent. (1470 degrees Fahr.), Fig. 5(c) and (d). The higher quenching temperature required for No. 95 is due not only

Table II
Compositions of Steels Tested

Steel No.	Composition Per Cent C	Mn	Si	Al	P	S	Residue Analysis MnO + MnS	SiO_2	Al_2O_3
				Commercial Steels					
94	.43	.79	.17	.004	.032	.030	.068	.025	.001
95	.37	.50	.20	.010	.018	.035	.040	050	.014
99	.52	.55	.22	.003	.040	.028	.064	.024	.031
103	.42	.61	.19	.008	.016	.026	.063	.017	.004
104	.31	.44	.17	.000	.010	.027	.049	.019	.002
105	.40	.73	.07	.006	.022	.025	.083	.012	.002
145	.48	.74	.21	.020	.022	.030	....	....	....
				Induction Furnace Melts, no Aluminum Addition					
2949	.31	.74	.09	.000	.009	.025	.069	.016	.026
2973	.41	.73	.21	....	....	....	.084	.008	.020
2974	.36	.72	.41	....	....	....	.082	.004	.023
2975	.44	.76	.90	....	....	....	.076	.027	.023
2978	.69	.75	.17	....	....	....	....	....	....
2977	.70	.76	.20	....	....	....	....	....	....
2976	.70	.66	.10	.011	(1.0% Co)		.061	.055	.042
				Induction Furnace Melts with Aluminum Addition					
2970	.41	.66	.10	.007	.000	.000	.064	.019	.006
2971	.39	.72	.11	.024	....	....	.069	.006	.035
2982	.42	.67	.13	.063	....	....	....	....	....
2972	.38	.73	.16	.077	....	....	.070	.011	.094
2981	.50	.82	.17	.118	....	....	....	....	....
2950	.48	.74	.15	.200	.010	.028	.071	.015	.098
2951	.39	.75	.16	.520	.010	.029	.062	.006	.100
2952	.39	.75	.15	.950	.009	.026	.075	.017	.118
2984	.41	.57	.09	.087	....	....	....	....	....
2983	.42	.74	.18	.070	....	....	....	....	....
2985	.39	1.01	.18	.084	....	....	....	....	....
2979	.63	.82	.24	(0.35% V)		....	....	....	....
				Hydrogen Purified Electrolytic Iron Alloys					
A128	.39	.003	.03	.000	.009	.009	.001	.014	.001
A130	.27	.003	.06	....	.008	.009	.001	.025	.003
A131	.97	.002	.05	....	.008	.008	.001	.016	.001
A138	.47	.690	.05	....	.016	.011	.048	.015	.003
A140	.53	.720	.25	....	.015	.012	.043	.032	.009
A158	.56	.003	.03	....	.010	.008	....	....	....
A174	.51	.790	.29	.030	....	....	....	....	....
A175	.49	.700	.25	(0.22% V)		....	....	....	....

to its lower carbon content, but also to a much higher content of visible nonmetallic inclusions, Fig. 5(e) and (f), with attendant local impoverishment of carbon content.

The expedient already adopted of plotting impact resistance against hardness offers also a possibility of comparing different compositions of steel. The curves for steels Nos. 94 and 95 as quenched from 830 degrees Cent. (1525 degrees Fahr.) are so plotted in Fig. 6 and show clearly the superior toughness of the latter steel. To minimize the number of tests required for comparing different compositions in this manner, the remaining steels were compared with the two just described, using only one tempering temperature, 300 degrees Cent. (570 degrees Fahr.). Hence, if the steel compared is different in carbon content from the standard, comparison at the same hardness value implies difference in tempering temperature.

The carbon content range, however, is small, so no material error is introduced by so doing.

Plotting observations on induction furnace steels of different carbon contents in Fig. 6, one may see that the effects of carbon variations within moderate limits are roughly equivalent to varia-

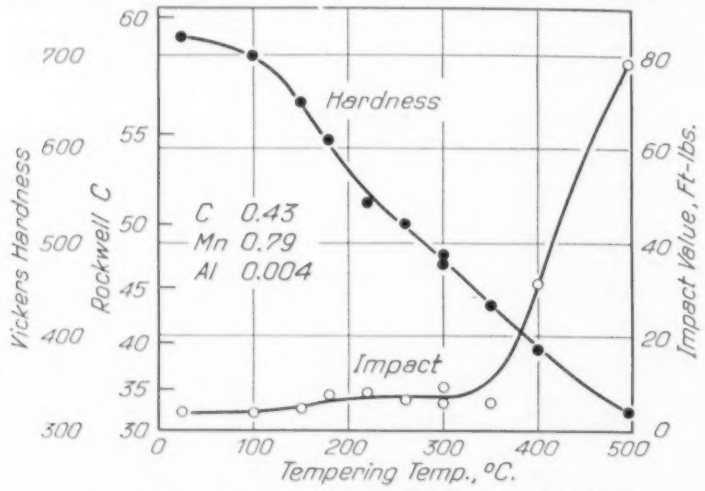


Fig. 2—Variation in Hardness and Impact Value of Steel No. 94 with Tempering Temperature. Specimens Quenched From 800 Degrees Cent. (1470 Degrees Fahr.).

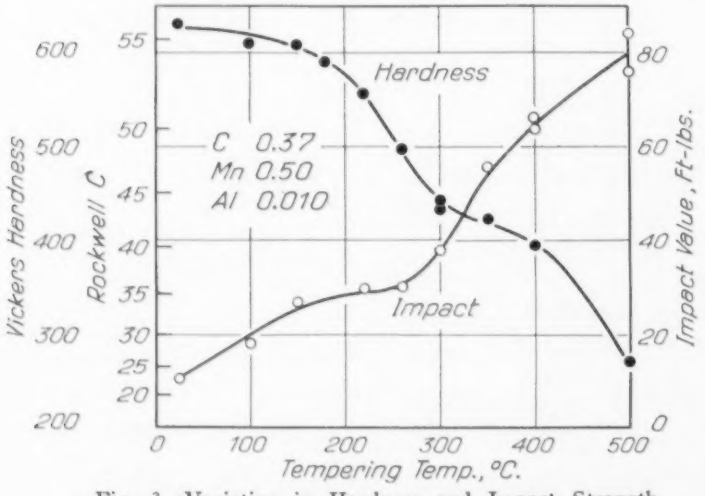


Fig. 3—Variation in Hardness and Impact Strength of Steel No. 95 with Tempering Temperature. Specimens Quenched from 830 Degrees Cent. (1525 Degrees Fahr.).

tion in tempering temperature. That is to say, both carbon content and tempering temperature are secondary variables so far as the comparison of steels over 400 Vickers hardness which differ only moderately in carbon contents is concerned. This fact justifies the basis already chosen for comparison of compositions.

Composition Factors

The content difference in the elements responsible for the striking difference in impact resistance between steels Nos. 94 and 95 is of such small magnitude that its analytical determination may be questionable. Fig. 6, however, indicates clearly the element probably responsible, namely, aluminum. Every individual observation on

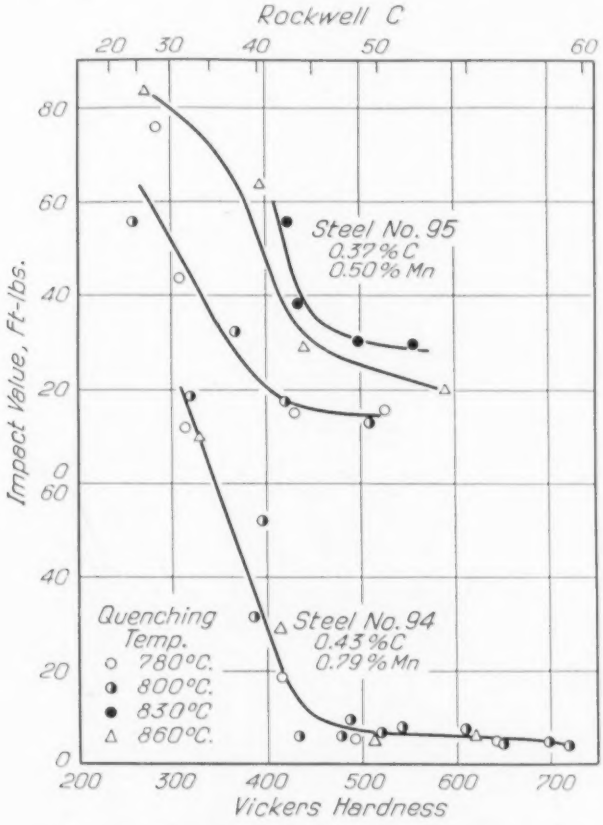


Fig. 4—Effect of Quenching Temperature on Impact Values of Steels Nos. 94 and 95 as Tempered at 25 to 400 Degrees Cent. (750 Degrees Fahr.).

laboratory melts deoxidized with 0.06 per cent or more aluminum (and also one to which vanadium was added) falls on the curve for steel No. 95. Other steels of the same charge, excepting that no aluminum or vanadium were added, give properties close to the curve of the low impact steel, No. 94, or values intermediate between these. Silicon under some conditions has a similar effect to aluminum and this may account for the single intermediate value.

The preceding observations suggest that aluminum was used in the deoxidation of steel No. 95 and its analysis supports this view.

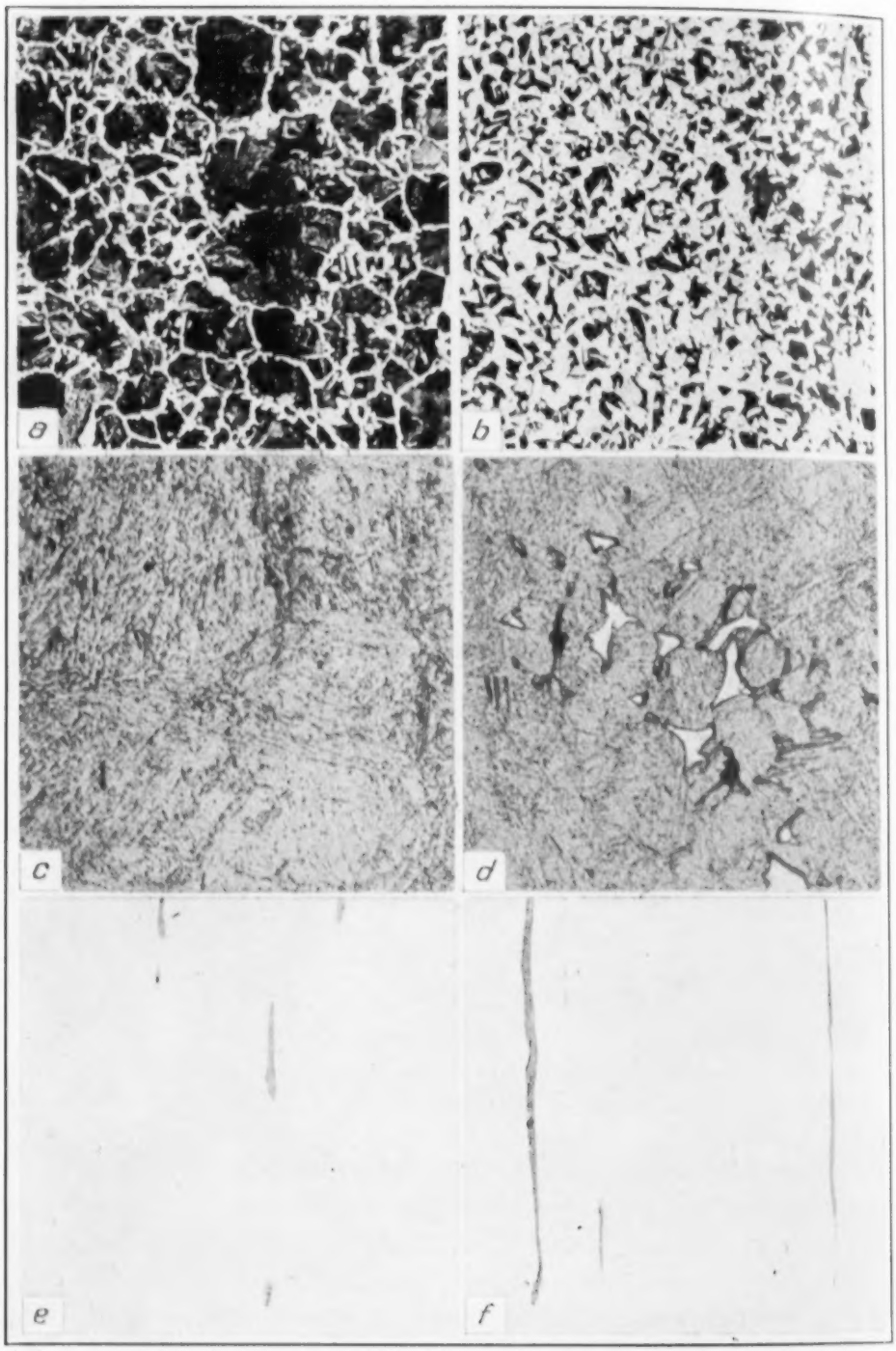


Fig. 5—Microstructure of Steels Nos. 94 (Left) and 95 (Right) Compared.
(a) and (b) as Air-cooled from 830 Degrees Cent (1525 Degrees Fahr.). × 100.
(c) and (d) as Quenched and Tempered at 180 Degrees Cent. (355 Degrees Fahr.). × 500. (c) Quenched from 780 Degrees Cent. and (d) from 800 Degrees Cent.
(e) and (f) Unetched to Show Inclusions. × 100.

Steel No. 94, on the other hand, was made under skilled observation and is known to have been well deoxidized without the addition of aluminum. Whether or not aluminum was used in the making of No. 95, it is certain that aluminum has a dominant effect on the impact resistance of hardened steel.

The laboratory melts were eminently satisfactory for determining the effects of aluminum, since the melts made with normal addi-

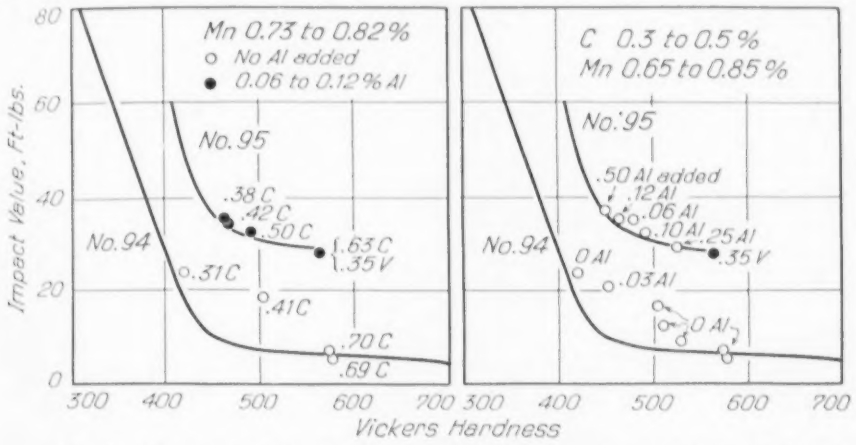


Fig. 6—Effect of Carbon Content on Impact Resistance of Laboratory-Melted Carbon Steels Quenched from 830 Degrees Cent. (1525 Degrees Fahr.) and Tempered at 300 Degrees Cent. (570 Degrees Fahr.).

Fig. 7—Effect of Aluminum Additions on Impact Resistance of Laboratory-Melted Carbon Steels Quenched From 830 Degrees Cent. (1525 Degrees Fahr.) and Tempered at 300 Degrees Cent. (570 Degrees Fahr.).

tions of the deoxidizers manganese and silicon only were consistently low in impact strength. Fig. 7 shows the effects of increasing aluminum additions made to melts of, with a few exceptions, precisely the same charge and deoxidized with manganese and silicon prior to the aluminum addition. It is clear that an addition of 0.06 per cent suffices to impart the upper level of impact resistance found in these steels.

Other elements than aluminum also influence impact resistance, but were not studied in detail. Some observations, however, are given in Fig. 8. Silicon apparently has an effect in the same direction as aluminum. In steels given a final addition of aluminum, increase in manganese content is definitely beneficial to impact resistance.

The observations so far given provide a basis from which to judge the impact resistance of fully hardened carbon steels tempered to between 400 and 600 Vickers. This data is put into convenient shape for use as a standard of comparison in Fig. 9 and has been

so used to establish deficiency in impact resistance of heat treated commercial steels that failed prematurely. It is used also in Table III to give the impact resistance of all the steels tested relative to a standard, the upper curve of Fig. 9.

Although laboratory melts made without the addition of aluminum or an equally effective deoxidizer were consistently low in impact strength, the same is not true of open-hearth steels. Thus steels Nos. 103, 104 and 105 gave a high impact value, Table III, though made without the addition of deoxidizers other than manganese and silicon. Only one steel of the group of four so made, No. 94, was low in impact strength. Additions of aluminum or equally active deoxidizers to open-hearth steels do not appear, therefore, to be essential for the attainment of high impact resistance. It is improbable, however, that other characteristics associated with high impact resistance have attained full maturity without such additions, a matter which current work will undoubtedly clarify.

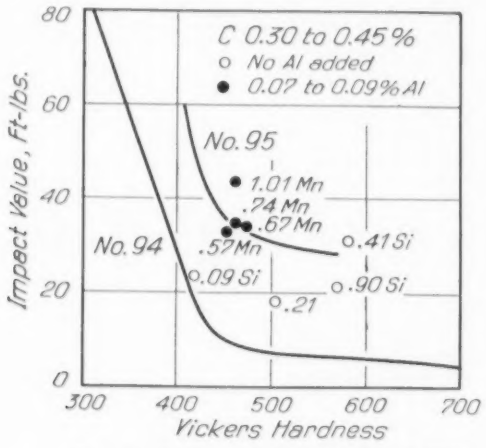


Fig. 8—Effect of Manganese and Silicon on Impact Resistance of Laboratory-Melted Carbon Steels Quenched from 830 Degrees Cent. (1525 Degrees Fahr.) and Tempered at 300 Degrees Cent. (570 Degrees Fahr.).

Grain-Size

Low impact resistance is associated with coarse grain-size in hardened steels and high impact resistance with fine grain-size. Clearly defined evidence thereof is difficult to obtain, however, in fully hardened steels. Thus the microstructure of martensite, Fig. 5(c) and (d), does not distinguish the large grain-size difference

between steels Nos. 94 and 95 disclosed by the ferrite network as normalized, Fig. 5(a) and (b). Since most of the steels under study were hypoeutectoid and, therefore, expelled ferrite to the austenite grain boundaries on air-cooling, this simple means for revealing austenite grain-size at time of quenching was used.

Pilot specimens of appropriate steels were cooled in air from the quenching temperature of the impact test specimens and their

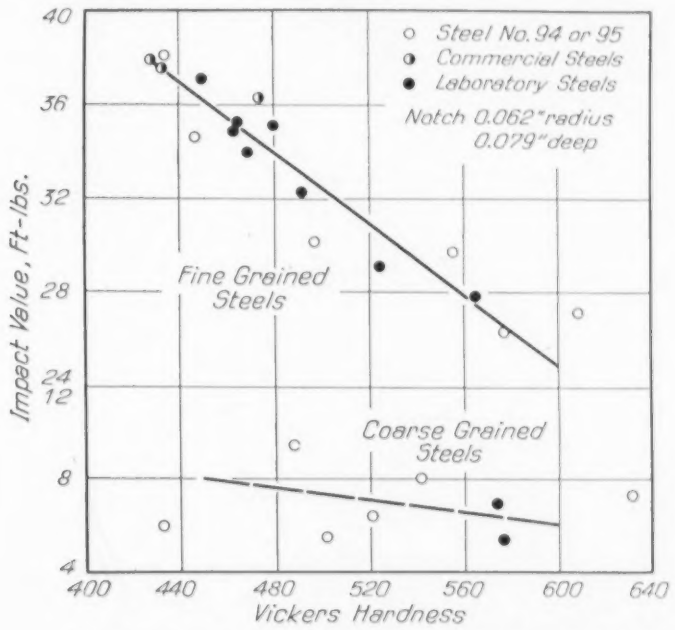


Fig. 9—Impact Values of Carbon Steels Quenched from 830 Degrees Cent. (1525 Degrees Fahr.) and Tempered at 150 to 300 Degrees Cent. (300-570 Degrees Fahr.) Plotted Against Hardness.

microstructure recorded in Figs. 10 and 11. As anticipated, the steels having a high impact resistance as hardened had a fine austenitic grain-size at time of quenching and those having a low impact resistance had a coarse austenitic grain-size. Grain-size effects, however, are not quite so simple as this conclusion based on average grain-size indicates.

The term grain-size usually implies the mean size of a large number of grains as seen on a plane section. Even if the grains cut are all of the same volume, the size of individual grains must vary uniformly between a maximum representing the true grain-size and zero because of varying distance between center of grain volume and the plane exposed. For a given mean grain-size, however, we can also have a mixture of grains of large volume and of small vol-

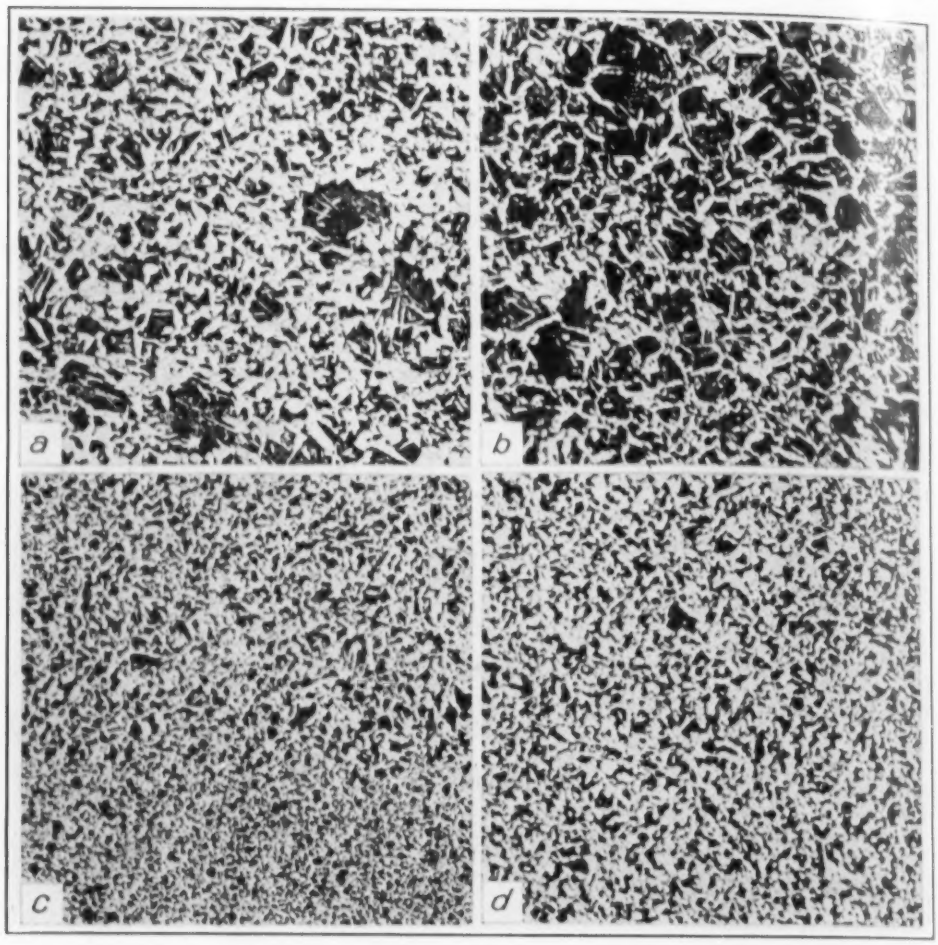


Fig. 10—Microstructure of Laboratory-melted Steels as Air-Cooled from 830 Degrees Cent. (1525 Degrees Fahr.) (30 Min.). Aluminum Contents are Amounts Added. Impact Values are those As-Hardened, Relative to Upper Curve of Fig. 9. × 100.
(a) Steel 2949, 0.00 Per Cent Aluminum, Impact Value —15
(b) Steel 2970, 0.03 Per Cent Aluminum, Impact Value —15
(c) Steel 2971, 0.06 Per Cent Aluminum, Impact Value + 1
(d) Steel 2972, 0.12 Per Cent Aluminum, Impact Value 0

ume. In this case the section would show "grain contrast" as in Fig. 5(a), a condition described by Grossmann (11) as "exaggerated grain growth."

Grain contrast is particularly prominent in grains of mean size intermediate between the fine structure of normalized aluminum, treated steels and the very coarse structure obtained by air-cooling from 1000 degrees Cent. (1830 degrees Fahr.) or higher. This is due to excessive rate of growth of a few individual grains. Grain contrast, however, diminishes again as the finer grains approach the size of the first large grains to form.

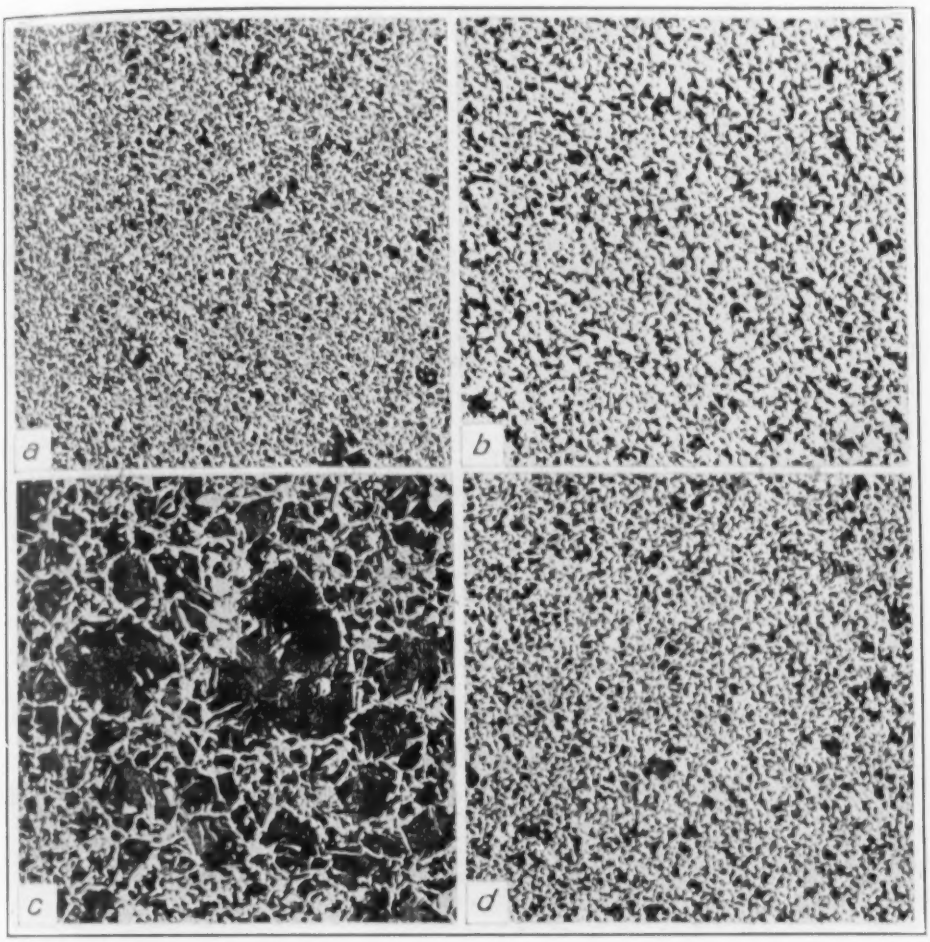


Fig. 11—Microstructure of Laboratory-Melted Steels as Air-Cooled from 830 Degrees Cent. (1525 Degrees Fahr.) (30 Min.). Aluminum and Silicon Contents are Amounts Added. Impact Values are those as Hardened Relative to Upper Curve of Fig. 9. × 100.

(a) Steel 2950, 0.25 Per Cent Aluminum, Impact Value — 1
(b) Steel 2951, 0.50 Per Cent Aluminum, Impact Value + 1
(c) Steel 2973, 0.30 Per Cent Silicon, Impact Value —14
(d) Steel 2975, 1.00 Per Cent Silicon, Impact Value — 6

Excessive grain contrast introduces the question: what fraction of the plane area must become coarse-grained before a fine-grained steel loses that part of its impact resistance attributable to fine grain-size? Sufficient work has been done on the impact resistance of pearlitic steels to disclose the answer to this question were tests on such steels significant. Lightner and Herty (12) found, however, that in pearlitic hypoeutectoid steels grain-size is only a secondary factor which acts indirectly through the basic factor, ferrite content, to determine impact resistance. With a given cooling rate, the amount of ferrite is lower the larger the grain-size, but for the same

ferrite content, grain-size has no detectable effect. Accordingly no conclusion on this point can be drawn from their work. It remains, therefore, for a study of grain-size effects in hardened steels to disclose a solution to this problem. A solution is necessary to express accurately the relation between grain-size and impact resistance.

Impact Fractures

Two types of notched-bar impact fractures are recognized and should be distinguished according to a report of A. S. T. M. Committee E-1. (13) One is the tensional type characterized by a granular fracture, little deformation of adjacent metal and a low value of energy absorbed. In the other case rupture follows shearing planes, the fracture may be fibrous and the energy absorbed is relatively high. As a contribution to the study of these fractures, photographs were taken of a number of representative specimens, Figs. 12, 13 and 14.

Purely tensional fractures were observed only in a few cases when the impact resistance was under 10 foot-pounds. An exception, however, was the extremely coarse-grained A-130 of Fig. 14 which had such a fracture at 15 foot-pounds. No fractures were entirely of the shearing type. The ultimate rupture near the center of the specimen even in the normalized specimens appears granular. Hence the major portion of the fractures were of a duplex type, showing shear failure over a symmetrical area near edges and tensional failure inside this area. The ratio of these areas, of course, varies greatly with the impact resistance, but is certainly not a constant for a given impact value.

At least three magnitudes in grain-size can be recognized from the granular part of the fractures. The pure iron-carbon alloy A-130 is coarsest, No. 94 steel is intermediate and No. 95 as well as the aluminum deoxidized steels are fine-grained. Grain-size is evidently a dominant factor determining the hardness at which transition from a granular to a partially fibrous fracture occurs. Thus the fine-grained steel, No. 95, has a partially fibrous fracture up to 610 Vickers at least, while the highest hardness at which the coarse-grained steel is partially fibrous is 395 Vickers.

This characteristic, the hardness at which the fracture shifts from granular to semi-fibrous, is a highly important index of the characteristics of the steel and may serve well to distinguish between

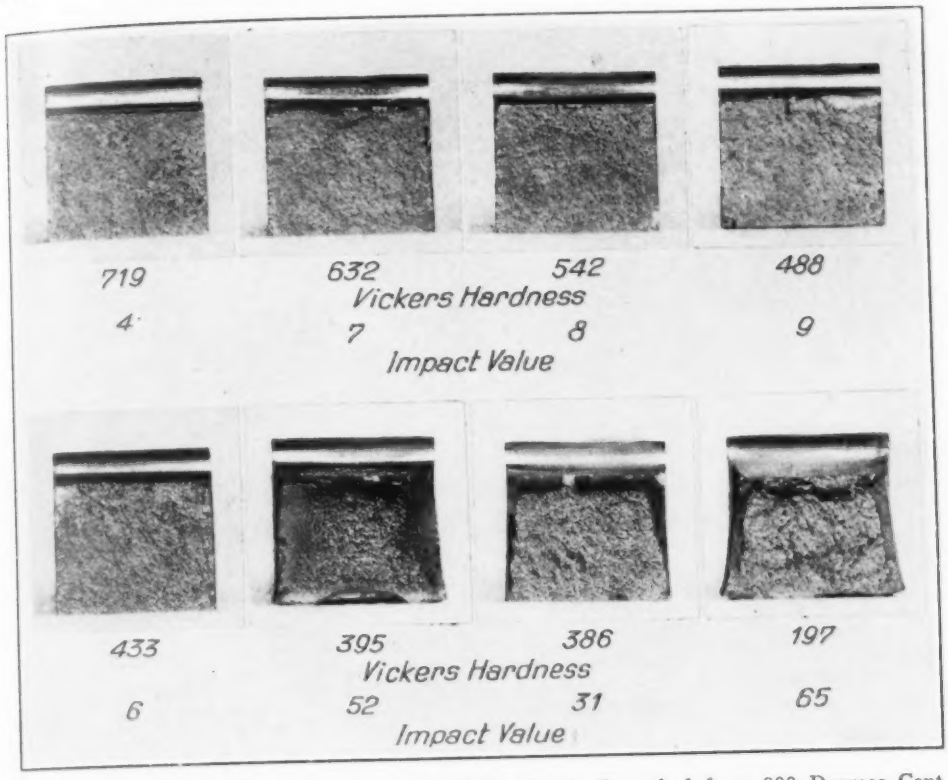


Fig. 12—Impact Test Fractures of Steel No. 94 As-Quenched from 800 Degrees Cent. (1470 Degrees Fahr.) and Tempered at 25 to 400 Degrees Cent. (750 Degrees Fahr.) Except Specimen of 197 Vickers which was Normalized. Magnification × 2.

inherently fine- and coarse-grained steels. This "transition hardness" is at least 600 Vickers for an aluminum treated steel. Higher carbon contents than 0.4 per cent would be required to obtain the necessary hardness to determine this point definitely in aluminum treated steels.

The transition in the character of the fracture occurs within such a very narrow range of hardness that it amounts almost to a discontinuity (see Fig. 2) and thus is easily recognized. It undoubtedly also depends on several other factors than those studied here, for example, size of specimen, shape of notch and temperature of test. Though variation of these factors may throw the transition hardness to a higher or lower value, the fact remains that the test conditions employed here serve well to distinguish clearly between two significantly different types of carbon steels; and that a certain degree of notch toughness and other desirable properties associated therewith are available to higher hardness values in inherently fine-grained than in coarse-grained steels.

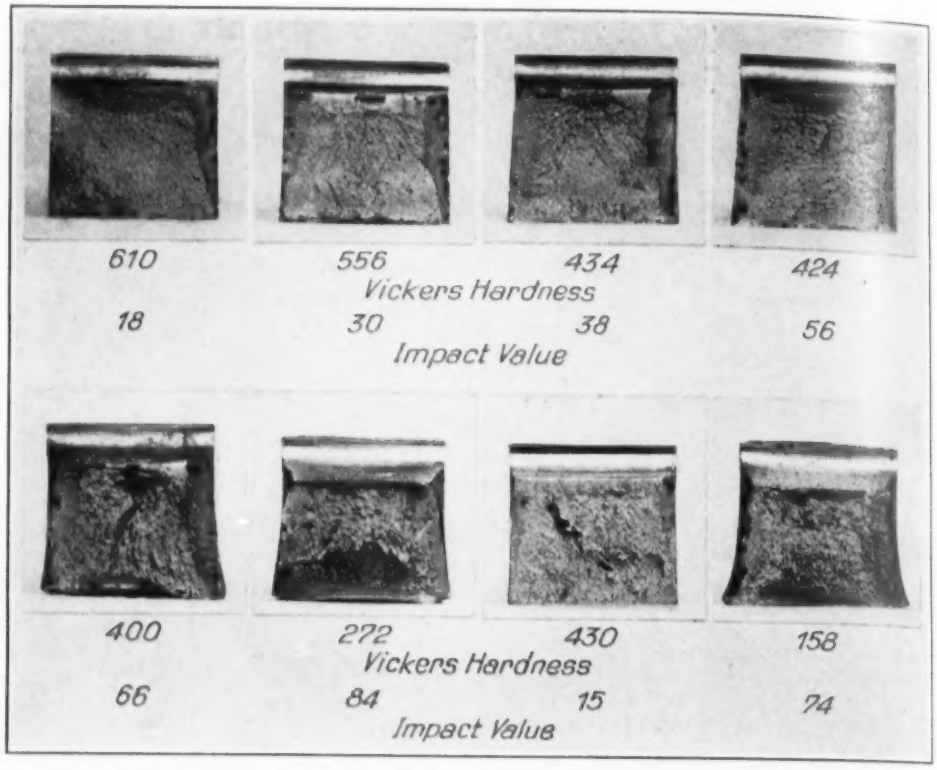


Fig. 13—Impact Test Fractures of Steel No. 95 As-Quenched from 830 Degrees Cent. (1525 Degrees Fahr.) and Tempered at 100 to 500 Degrees Cent. (930 Degrees Fahr.), Except 272 and 430 Vickers Specimens which were Quenched from 860 and 780 Degrees Cent. (1580-1435 Degrees Fahr.) Respectively and 158 Vickers which was Normalized. Magnification × 2.

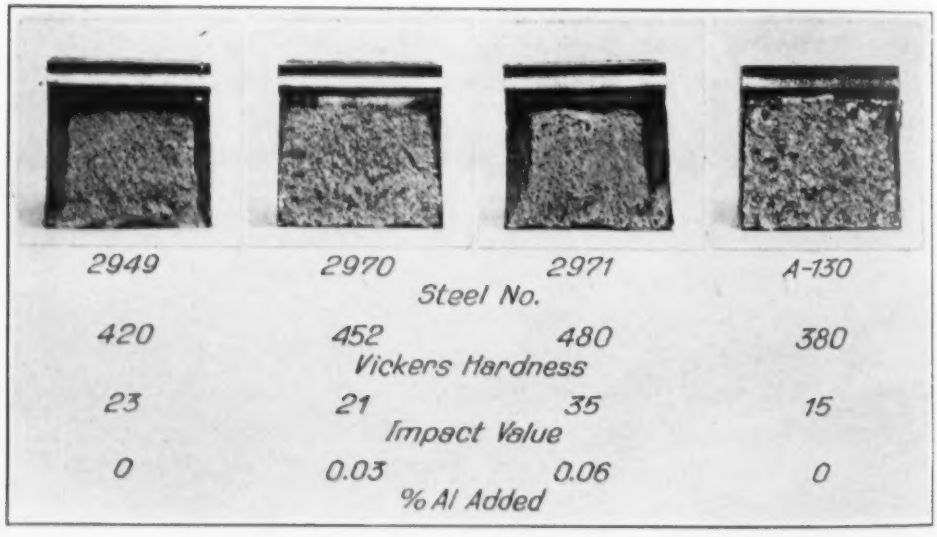


Fig. 14—Impact Test Fractures Showing Effects of Aluminum-Treatment. Specimens Quenched from 830 Degrees Cent. (1525 Degrees Fahr.) and Tempered at 300 Degrees Cent. (570 Degrees Fahr.). Magnification × 2.

Table III
Properties of Carbon Steels Tested

Steel No.	Composition Per Cent C	Mn	Si	Al	X	Vickers Hardness As Quenched	Tempered 300°C	Impact Value Tempered 300°C	Relative to Standard
					Commercial Steels				
94	.43	.79	.17	.004	.032P	680	480	7.0	−26
95	.37	.50	.20	.010	.018P	655	440	38.0	+ 1
99	.52	.55	.22	.003	.040P	...	522	12.6	−18
103	.42	.61	.19	.008	.016P	...	474	36.3	+ 2
104	.31	.44	.17	.000	.010P	...	428	37.9	0
105	.40	.73	.07	.006	.022P	...	433	37.5	0
					Induction Furnace Steels, no Aluminum Addition				
2949	.31	.74	.09	.000	....	620	420	23.4	−15
2973	.41	.73	.21	....	....	800	505	18.2	−14
2974	.36	.72	.41	....	....	685	581	30.9	+ 5
2975	.44	.76	.90	....	....	725	570	21.6	− 6
2978	.69	.75	.17	....	....	871	577	5.4	−21
2977	.70	.76	.20	....	....	850	574	6.9	−20
2976	.70	.66	.10	.011	1.0Co	935	571	12.1	−15
					Induction Furnace Steels with Aluminum Additions				
2970	.41	.66	.10	.007	.03*	735	452	20.6	−15
2971	.39	.72	.11	.024	.06*	695	480	35.1	+ 1
2982	.42	.67	.13	.063	.10*	...	469	33.9	− 1
2972	.38	.73	.16	.077	.12*	705	465	35.2	0
2981	.50	.82	.17	.118	.10*	752	492	32.2	− 1
2950	.48	.74	.15	.200	.25*	760	525	29.1	− 1
2951	.39	.75	.16	.520	.50*	650	450	37.1	+ 1
2979	.63	.82	.24	....	.35V	880	565	27.8	0
2984	.41	.57	.09	.087	.10*	...	453	32.7	− 3
2983	.42	.74	.18	.070	.10*	...	463	34.8	0
2985	.39	1.01	.18	.084	.10*	...	463	43.6	+ 8
					Hydrogen Purified Electrolytic Iron Alloys				
					P+S				
A130	.27	.003	.06	.000	.017	...	380	15.4	....
A138	.47	.690	.05	....	.027	...	510	12.1	−20
A140	.53	.720	.25	....	.027	...	530	9.1	−21

NOTE—No. 2981 cast immediately after Aluminum addition, No. 2982 held 10 minutes after aluminum addition before casting and No. 2983 held 30 minutes.

*Per cent Aluminum added.

Action of Aluminum

Aluminum additions to liquid steel offer an outstanding example of large effects produced by small doses of an alloying element. After the metallic aluminum content reaches about 0.01 per cent, further additions up to 0.5 per cent at least have no appreciable effect on impact resistance, Table III. Such a relation is not characteristic of solid solution alloying effects, so one must seek elsewhere an explanation of the beneficent effect of aluminum.

It has been suggested that the function of aluminum in this regard is to produce by reaction with the oxygen in solution in liquid steel a shower of sub-microscopic alumina particles. (5) Certainly alumina is formed, for it is recovered in easily measurable quanti-

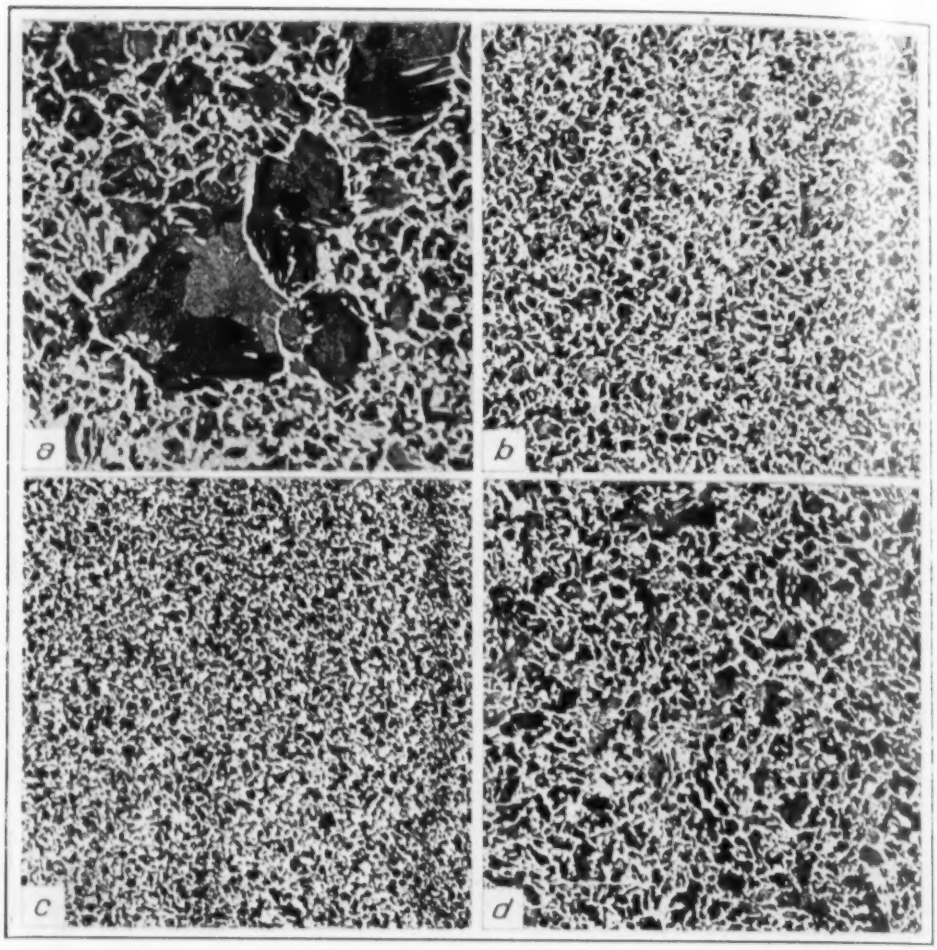


Fig. 15—Microstructure of Steels Prepared from Hydrogen-Purified Iron Melted and Frozen Under Hydrogen, Compared with a Fine-Grained Commercial Steel. Specimens Air-Cooled from 830 Degrees Cent. (1525 Degrees Fahr.) (30 Minutes). × 100.
(a) Steel No. A-140, Manganese and Silicon only Deoxidizing Elements added.
(b) Steel No. A-174, 0.10 Per Cent Aluminum added after Manganese and Silicon.
(c) Steel No. A-175, 0.25 Per Cent Vanadium added after Manganese and Silicon.
(d) Fine-Grained Commercial Steel No. 145.

ties by suitable extraction methods. Being a suspension of much lower density one would expect it to rise to the surface of the metal or to flux with other nonmetallic matter if the bath is held molten for a sufficient time. Experiment, however, did not reveal any such effect. Steels Nos. 2982 and 2983 of Table III were held molten 10 and 30 minutes respectively after an aluminum addition of 0.10 per cent instead of less than one minute as was otherwise the practice. In neither case was the impact value below normal. It should be noted, however, that the melt was quiescent during the holding period for the high frequency magnetic flux did not penetrate

through the graphite crucible shell sufficiently to cause observable stirring. Hence it may be concluded that the alumina particles do not flux or rise rapidly in a quiet melt.

Still assuming that aluminum shows its characteristic effects through colloidal alumina, it follows that addition of aluminum to an oxygen-free melt would not produce the same effect, for lacking oxygen no alumina can form. Substantially oxygen free steels were produced by hydrogen purification of electrolytic iron stock, series A of Table II. They were melted and frozen under hydrogen also. That steels so prepared are extremely low in oxygen content was proved by analysis for alumina after adding over 1 per cent aluminum. Herty (14) has shown that with such an aluminum addition all of the oxygen is recovered as alumina. The alumina content of this alloy, nevertheless, was only 0.005 per cent.

Additions of 0.10 per cent aluminum and 0.25 per cent vanadium were made to pure steels melted as just described, steels Nos. A174 and A175 of Table II. Specimens of these steels and comparison steels without such additions were cooled in air from 830 degrees Cent. and their microstructure recorded in Fig. 15. The steels to which aluminum and vanadium were added are decidedly finer grained than the others made from hydrogen purified iron and melted and frozen under hydrogen. Since it can be argued that oxygen was not completely eliminated this experiment also was not conclusive. It may be concluded, however, that if alumina is the active ingredient, an extremely small quantity is sufficient to produce the characteristic effect.

Metallographic study of highly purified iron to which individual additions of the several constituents of ordinary steels, such as oxygen, manganese, and so on, have been added, have much else of interest to offer that can be barely mentioned here. When heated at temperatures where grain growth becomes pronounced, they show an exceptional degree of grain-size contrast, Fig. 15. When qeunched as 1-inch rounds in the submerged water spray, these pure alloys fail to harden in the sense of acquiring a continuous martensitic shell. Only the exceptionally large grains harden and they harden whether at the center or near the surface. Even with normal contents of manganese and silicon penetration of hardening is slight. Means are available, however, for overcoming these objectionable characteristics.

The highly purified steels were particularly free from nonmetal-

lic matter and ferrite concentrations. Ferrite solution was obtained at lower temperature than in commercial steels with corresponding grain-size. Directional effects appeared to be nearly absent. These then are advantages to be gained by further progress in the purification of commercial steels, which may appear in the "steels of tomorrow."

On the basis of these observations it may be suggested, furthermore, that the commercial steel here used as a glaring example of low impact strength, No. 94, is in reality an example of higher purity than the average in open hearth steels and is a tribute to the achievements of Dr. Herty in the improvement of steel melting practice.

Summary and Conclusions

A review of current opinion on the significance of the notched-bar impact test suggests that this test has exceptional utility when applied to quenched steels in the martensitic or troostitic condition. Impact, hardness and metallographic tests were made on a variety of carbon steels quenched in a submerged water spray to secure uniform deep hardening. Using a Charpy impact specimen with a special notch produced by grinding and lapping after heat treatment, consistent impact test results were obtained.

Observations of variation in impact values of hardened carbon steels with changes in several factors permitted the following conclusions:

(1) There is no pronounced maximum in impact resistance of specimens loaded by bending for tempering temperatures in the neighborhood of 180 degrees Cent. as is reported by Luerssen and Greene from torsional impact tests.

(2) Free ferrite appears to affect impact resistance adversely.

(3) The impact resistance of hardened carbon steels of a particular type as to grain-size is a function of hardness that is apparently independent of carbon content, when their hardnesses are maintained between 400 and 600 Vickers by tempering.

(4) Nearly all observations on a variety of hardened steels fell on one or the other of two curves of impact value plotted against hardness.

(5) Laboratory melted steels deoxidized only with manganese and silicon in normal amounts fall on the curve of low impact values

and those to which a final addition of at least 0.06 per cent aluminum were added fall on the curve of high values.

(6) In larger additions than is necessary with aluminum, vanadium has the same effect on impact resistance.

(7) Certain commercial steels known to have been made without the addition of aluminum or vanadium gave high impact values.

(8) The impact value of aluminum treated laboratory melts increases materially as the manganese content varied upward from 0.6 to 1.0 per cent.

Observations of grain structure and impact fractures permitted further conclusions as follows:

(1) Aluminum treated steels melted in the laboratory under careful control are fine-grained as well as high in impact resistance while those made without the aluminum or an equivalent addition are coarse-grained and low in impact resistance.

(2) Wholly granular impact fractures were observed in only a few instances and corresponded with minimum impact resistance. Granular fractures were obtained in coarse-grained steel at hardness values over 400 Vickers, but not in fine-grained steel up to 600 Vickers.

(3) Steels made from highly purified iron and known to be extremely low in oxygen content were fine-grained after aluminum and vanadium additions.

(4) Other alloys made from purified iron without the addition of aluminum or vanadium develop extreme grain contrast on heating slightly above the end of Ac_3.

(5) Penetration of hardening in the purified alloys is very shallow except that isolated grains of macro dimensions are martensitic at the center as well as near the edge of quenched 1-inch round bars.

Observations on the highly purified alloys further suggested that:

(1) Increased purity of commercial steels will entail poorer impact resistance in the hardened condition unless a compensating addition of aluminum or another element having an equivalent action is made.

(2) Higher purity in commercial steels will tend to eliminate directional effects and irregular distribution of ferrite, thereby permitting lower quenching temperatures than are now required.

ACKNOWLEDGMENTS

The writer was ably assisted in the experiments described herein by J. M. Kelly, who prepared the laboratory steels, and by Miss M. Ferguson, who did the metallographic work. H. H. Schmitt of the Materials and Process Engineering Department prepared the pure iron and steels made therefrom, using a procedure developed by P. H. Brace of the Research Laboratories. Residue extractions were made by C. W. Weltman of the Chemical Division and impact tests by T. F. Hengstenberg of the Mechanics Division, Westinghouse Research Laboratories.

The co-operation of Dr. C. H. Herty, Jr., who supplied materials and enlightening comments, was enjoyed throughout the course of this investigation.

References

1. N. N. Davidenkoff, *Transactions,* American Society of Mechanical Engineers, Vol. 56, 1934, p. 97.
2. R. H. Greaves, First Communications NIATM A:225 (1930).
3. T. W. Hardy, *Iron Age,* Vol. 122, 1928, p. 1557.
4. B. F. Shepherd, *Metal Progress,* Vol. 19, 1931, p. 51.
5. E. C. Bain, TRANSACTIONS, American Society for Steel Treating, Vol. 20, 1932, p. 385.
6. N. N. Sawin and E. Stachrowski, TRANSACTIONS, American Society for Steel Treating, Vol. 20, 1932, p. 249.
7. G. R. Fitterer, Bulletin No. 51, Mining and Metallurgical Advisory Board, Carnegie Institute of Technology.
8. C. H. Herty, Jr., C. F. Cristofer, M. W. Lightner and H. Freeman, Bulletin No. 58, Mining and Metallurgical Advisory Boards, Carnegie Institute of Technology.
9. H. Scott, TRANSACTIONS, American Society for Metals, 1934.
10. G. V. Luerssen and O. V. Greene, *Proceedings,* American Society for Testing Materials, Part II, Vol. 33, 1933, p. 315.
11. M. A. Grossmann, TRANSACTIONS, American Society for Steel Treating, Vol. 21, 1933, p. 1079.
12. M. W. Lightner and C. H. Herty, Jr., Bulletin No. 59, Mining and Metallurgical Advisory Boards, Carnegie Institute of Technology.
13. Report of Committee E-1: *Proceedings,* American Society for Testing Materials, Part I, Vol. 30, 1930, p. 909.
14. C. H. Herty, Jr., J. M. Gaines, H. Freeman and M. W. Lightner, *Transactions,* American Institute of Mining and Metallurgical Engineers. Iron and Steel Division, 1930, p. 28.

DISCUSSION

Written Discussion: By O. V. Greene and G. L. Luerssen, Carpenter Steel Co., Reading, Pa.

Mr. Scott draws some interesting comparisons between the results obtained on his notched bars and those obtained by the torsion impact method presented

by Mr. Luerssen and myself before this Society last year,[1] pointing out that the peak of impact resistance at a draw of approximately 350 degrees Fahr. found by the torsion method is absent in the notch test. We wish to call attention to the fact that the steels discussed by Mr. Scott are in general quite different from those which we covered in that paper. Mr. Scott gives a typical example of two steels containing 0.37 and 0.43 per cent carbon, while ours were confined to a range of approximately 0.90 to 1.10 per cent carbon.

The lower carbon ranges were purposely avoided in our work since we felt that they could be studied successfully by the notch methods regularly used. These methods, however, were found useless on the tool steels. In an earlier paper before the A.S.T.M.[2] for instance we showed comparative results by the torsion method and the Izod using the standard notch on a 1.06 per cent carbon steel. It was noted at that time that comparative tests were also made on the Izod machine using a notch having a depth of 0.0805 inch and a radius of $\frac{1}{16}$ inch, and it is interesting that this notch agrees quite closely with that used by Mr. Scott on the Charpy bars. The results obtained with this notch were so close to those of the standard Izod notch that only the latter were reported. These showed a linear increase from 2 foot pounds with no draw to 13 foot pounds at a drawing temperature of 800 degrees Fahr.

That Mr. Scott did not observe pronounced maximum points would be expected for two reasons.

First, a full martensitic structure appears to be necessary even by the torsion impact method to secure a pronounced peak at a drawing temperature of approximately 350 degrees Fahr., the impact resistance falling off when troostite begins to develop at higher drawings. Steels No. 94 and No. 95 taken as typical by Mr. Scott are of such a carbon content that a troosto-martensitic structure would be developed upon quenching and it would not be expected therefore that a definite peak would be observed. In fact we have evidence that the torsion impact test would give very much the same type of curve on this type of steel as those shown by Mr. Scott on the notched impact tests.

Second, the presence of the notch would naturally greatly reduce the impact resistance of the full martensitic structure due to concentration of stresses in this harder and relatively nonductile constituent. The issue here appears to be whether the true impact value of martensite is shown by the notched test which concentrates the stresses at one point, or an un-notched test which distributed the stresses. In any one specific case the question might well be settled on the basis of whether the tool or part in question contains sharp corners or notches at vital points, and of such design that the stresses occur through such a notched section, or whether the vital parts of such tool are free from notches or recesses which would concentrate stresses. The danger of such notches is well known among tool makers and hardeners and they are of course avoided wherever possible. It is consequently our feeling that the true impact value of martensite is obtainable only on an un-notched specimen in which the distribution of stress is uniform, and that the introduction of any arbitrary notch will

---

[1] G. V. Luerssen and O. V. Greene, "Torsion Impact Properties of Hardened Carbon Tool Steel," TRANSACTIONS, American Society for Metals, Vol. 22, 1934, p. 311.

[2] G. V. Luerssen and O. V. Greene, "The Torsion Impact Test," American Society for Testing Materials, Vol. 33, 1933, p. 315.

greatly reduce the value of the test in determining fundamental information. Consequently we feel that Mr. Scott's statement that the impact resistance of martensitic and troostitic steels is of much lower magnitude than that of the softer steels, might well be modified to read "notch impact resistance," and the conclusions upon the impact resistance of martensitic steels cannot safely be based upon tests made on 0.40 per cent carbon steel.

Mr. Scott's specimens were evidently prepared by first heat treating the full specimen, then subsequently grinding the notch. It would be interesting to know if any hardness profiles were made to determine whether the hardnesses at the base of the notch were equal to the surface hardnesses, and also whether or not the structures were similar. Particularly on the higher carbon steels this is an important point since it is readily possible to have a difference in hardness between the surface and a point 0.079 inch below the surface.

On page 1146 Mr. Scott comments on the lower impact values secured on quenching at temperatures sufficiently low to produce free ferrite. This is not surprising and in our own work we have repeatedly found that mixed structure will impair the impact value.

On page 1152 Mr. Scott draws some interesting comparisons between his laboratory melts and commercial open-hearth steels, finding that the latter have a higher impact even though made without the addition of deoxidizers other than manganese and silicon. It is unfortunate that all of the electric furnace steels tested have been made in the induction furnace. It would be interesting to have had comparisons also with arc electric melts.

We are entirely in accord with Mr. Scott in his conclusions that the fine-grained steels have a uniformly higher impact value than the coarse-grained steels at the same hardness. His analysis of the various impact fractures obtained is very logical and interesting.

We wish to commend Mr. Scott on a very valuable and carefully conducted investigation.

Written Discussion: By H. H. Ashdown, consulting metallurgist, Pittsburgh.

We have been very interested in reading Mr. Scott's paper for several reasons. Although apparently his original object was to show the advantages to be derived by the addition of small quantities of aluminum to the liquid steel before pouring, he is to be congratulated on his frank statements that the results obtained by laboratory methods may not be paralleled on a production scale.

All the small melts made in the laboratory crucible with and without aluminum additions produced the results anticipated. But as pointed out by Mr. Scott, on page 1152, certain melts made in the open-hearth furnace gave, without the addition of aluminum results equally as good as those to which aluminum had been added.

The addition of aluminum is unquestionably productive of fine grain structure and greater depth hardening properties during quenching. There are, however, the objectionable features of the residual Al_2O_3 causing elongated streak inclusions in the direction of subsequent working. These are detrimental, particularly when forgings are subject to transverse and radial test. Again they are more than objectionable in dies for drop forge work which are subjected to continuous shock stresses.

It is clearly stated that after the addition of aluminum, despite a comparatively long period of rest in the molten condition, the Al_2O_3 remained in suspension. Had impact tests been taken in a direction transverse to forging, inferior results would doubtless have been shown and these would have been very enlightening.

It has been our experience that a well-made open-hearth steel finally deoxidized in the ladle by the addition of crushed ferromanganese during the early stage of tapping, give results superior to those produced by the addition of aluminum, particularly when the final product is tested transversely and/or radially. Further, for many years there has been on the European market a material known as S.A.M. metal, an alloy of silicon, aluminum and manganese and this material has been used as a ladle deoxidizer giving results equal to those of aluminum additions with greatly reduced Al_2O_3 inclusions. It is possible Mr. Scott would find an alloy of this type advantageous when making additional experiments.

Aluminum when plunged into the liquid bath is rapidly oxidized on the surface with the formation of a very tough protective skin. We have seen on several occasions sticks of aluminum secured to a steel bar by stout high temperature melting wire, after being plunged into the bath (in a portion free from slag) and vigorously swayed in the bath, on withdrawal of the bar from the bath the aluminum still intact, and on breaking the tough skin the molten aluminum has run out.

It may therefore be that when aluminum is added to the ladle, the fall of the liquid stream is sufficient to break up the original form of the aluminum into small particles each strongly enveloped by oxide film and these are retained in suspension due to a slow process of diffusion reaction with the aluminum within the envelope.

Mr. Scott is to be congratulated further, on his choice of form of impact test piece although we see little advantage in departing from the standard Charpy test piece. The standard Izod impact test piece made from any material and in any condition is costly to prepare and in its final preparation there is always the risk of transverse marks (longitudinal to the fracture) however faint, being left at the root of the notch, adversely influencing the result of test.

With the Charpy test piece all work in the preparation of the notch is transverse to the direction of fracture and any fine surface irregularities which remain, influence the final result in a very minor degree.

Oral Discussion

G. C. RIEGEL:[3] In the main, I wish to commend Mr. Scott for focusing our attention on this method of dynamic testing of steel. It seems to me that there are several statements in the paper which, from our experience, are necessary both to affirm and disclaim. The author states that service failures have not been definitely related to Charpy impact characteristics of carbon steel in a harder condition than the sorbitic state. From seven years experience in Charpy impact testing of S.A.E. 1045 steel, running above 1000 mill heats, and from several different mills, it has been definitely determined that good Charpy im-

[3]Metallurgist, Caterpillar Tractor Co., Peoria, Ill.

pact characteristics at 600 and 800 degrees Fahr. tempering temperatures prevent service failures on tractor parts used in hardness ranges covering Rockwell C35 to 55. (Ref. December 3, 1931, Iron Age, p. 1426-29).

It must be agreed that ferrite affects impact resistance of carbon steel adversely but by the same token it must be concluded that factors controlling grain-size may be detrimental when they unduly hasten the transformation of austenite in treatment; in other words, inherently fine-grained carbon steel may harden so ineffectively as to give poor shock resistance.

The observation of the author of increased purity of the steels needs further comment. We are of the opinion that a standard for purity ought to be established in the manner similar to the grain-size standard of the A.S.T.M. If purity greater than shown in Fig. 5, marked "f" is indicated as being *undesirable* for hardened carbon steels of high dynamic qualities, then we are fearful that the inference is misleading, or that insufficient studies for the correlation of these factors have been brought out.

F. B. Foley:[4] I should like to start a general discussion concerning just what it is that causes fine grains and coarse grains in steel. There have been several references here to the fact that alumina might form nuclei of crystallization generally distributed throughout steel and other suggestions that the alumina particles form actual barriers beyond which the grains cannot grow. It may be that neither of these things is the cause of it. In the long run, we may find out that it is not a mechanical process at all.

A heat of steel that has very little silicon added makes a rimmed steel which, during solidification, throws off a great quantity of gas—it effervesces. When silicon is added the steel does not throw off gas, and the amount of silicon required to keep the gas from going off is in excess of about 0.10 per cent. This is silicon in solution in the metal. Its effect may be to increase the ability of the steel to retain gases in solution and not give them off. It may be that without the silicon present the gas escapes and that when silicon is added is retained in solution. Aluminum may have the same sort of action. The whole question of the mechanism by which small grain-size is produced by deoxidation methods in steel melting is very interesting, and it will make a good subject for discussion if we can get somebody to talk about it.

I have always pictured an effervescent heat of the type observed in ginger ale, of particles of CO, carrying with them a thin film of iron which oxidizes when it strikes the air. Of course this gas is formed by an interaction of carbon with oxygen, probably in the form of iron oxide. The iron oxide will also react with silicon to form SiO_2. When silicon is added to steel the metal is not killed until a certain amount of residual silicon is present.

The point is that just enough silicon to combine with the oxygen, leaving no silicon in solution, is not sufficient evidently to kill the metal. What is the action of the silicon in excess of this, which is in solution in the metal and which is necessary to produce killed steel? I have heard it suggested that this action is one of increasing the solubility of gases in the metal. If a large excess of aluminum is added, comparable with the excess of silicon commonly found necessary, I believe fine grain will not result.

---

[4]Metallurgist, Midvale Co., Nicetown, Philadelphia.

Conceivably colloidal particles of alumina may, by serving as nucleii of crystallization, cause the formation of small austenite crystals, but it is very difficult to imagine particles of alumina migrating through the austenitic crystals to their grain boundaries there to form a barrier.

Grains I think grow by one crystal robbing its neighboring crystal of its material atom by atom at grain boundaries. I do not believe that crystals grow by a coalescence of two crystals, one of them changing its orientation and then joining up to make a single crystal of two. The process of grain growth must be a process of the larger crystals robbing the smaller ones, and in this conception we might suppose that crystals have what may be called an intercrystalline force. Now that force would tend to draw the atoms at the boundary of adjacent crystals into the orientation of the larger crystal and thus, atom by atom, robbing it. This would go on until the crystals were all approximately of the same size and then it would slow up until the crystal growth promoting influence was sufficient to give one that was a little bit larger than the other one, sufficient force to again start robbing its neighbors more rapidly.

If numerous small crystallizing centers are present, particles of alumina for example, and they are very uniformly distributed, they might start crystallization in austenite and produce a rather uniform small grain-size, which would not grow rapidly because of the uniformity of the grains produced. Now if very few of them were present large and small grains might be produced and then grain growth would occur much more rapidly because the larger ones would eat up the little ones faster. In the case of a grain-growth barrier, we conceive of an insoluble oxide existing in the crystal which tends to push it out to grain boundaries and, once it gets there, it acts as a barrier. These are the only conceptions I have of a mechanical action of aluminum particles, but Mr. Scott suggests to us that possibly you do not require alumina at all, that it is merely the action of aluminum. Of course, that would lead to an entirely different conception.

R. S. DEAN:[5] Any steel-making process obviously involves a slag and a metal phase. In applying analytical methods to steel-making processes, we make of course an assumption that the reactions proceed homogeneously within the phases.

There are, however, pretty good reasons to believe that you may emulsify 999 parts of steel with one part of slag, and that you would get in that a steel melt that would be made up of the emulsion cells of a mixture in which the slag is the external phase but present in very small amounts. Now obviously those cells would have a strong determining effect on grain-size and the effect of small amounts of silicon and aluminum may be readily compared with the effect of small amounts of certain wetting agents. You can take a mixture of oil and water and add a few thousandths of a per cent of some wetting agent and emulsify the whole thing. You can also reverse that emulsion and change it over from an emulsion of water in oil to oil in water.

It strikes me that the effect of aluminum and some of these other things may be one of reversing the emulsion from a condition in which the slag is

[5]Chief engineer, metallurgical division, U. S. Bureau of Mines, Washington.

the external phase to one in which it is the internal phase. I have only a little experimental evidence available. If you take an iron oxide surrounded by a small amount of silica, and alumina, and if you reduce that you get metallic iron surrounded with silicon alumina in a typical sponge structure. If you heat that up to 1450 degrees Fahr., the silicon aluminum fluxes with the iron oxide and you have very small particles of iron surrounded with a very thin sheet of slag. You can put that through a set of rolls to take out any large amounts of slag, and you get a product which might be classed as fine wrought iron, which has one characteristic—if you melt it you never get that slag out of it because it is emulsified as the external and not the internal phase. The minute you deoxidize it then you reverse the emulsion and get the slag as the internal phase.

C. H. Herty, Jr.:[6] I just want to ask if it is his belief that silicon makes iron absorb oxygen. Do you really believe that?

Jerome Strauss:[7] Some questions are raised by Mr. Scott's mention of his observations on hydrogen-treated metals and the reactions between metals treated by hydrogen. Have we any reason to assume that oxygen and nitrogen are completely removed by treatment with hydrogen at high temperatures? Whether this question can or can not be answered, what is likely to be the effect of interaction between metals which immediately prior to their preparation were not in the form of oxides or, more specifically, what would result if such metals were prepared by the decomposition of organic compounds containing neither oxygen nor nitrogen? Here is a field for both speculation and experiment.

General Discussion

G. R. Brophy:[8] In all of the discussion this afternoon, it seems to me that oxygen or oxide particles have been assumed a necessity. It is my point of view that any solid particle of the correct dispersion and insoluble within the temperature range considered may act as a grain growth inhibitor, thereby rendering the steel fine-grained and abnormal.

Bain, in his experiment with aluminum-iron alloy, showed that the steel was normal when carburized in hydro-carbon atmosphere but was abnormal when carburized in a CO atmosphere as a result of the formation of Al_2O_3 particles.

If with hydrocarbon carburization we could produce an insoluble carbide properly dispersed, then fine grain and abnormality should result. This we did with a vanadium-iron alloy prepared from hydrogen purified material melted in hydrogen and subsequently carburized in an oxygen-free hydrocarbon atmosphere. The resulting product was highly abnormal and fine-grained. Thus we see that oxygen or oxides are not necessary for the production of fine-grained abnormal structures in steel.

L. L. Wyman:[9] There is another factor that also enters into this question, and it has its analogies in the nonferrous field, being familiar to quite a

---

[6]Research metallurgist, Bethlehem Steel Co., Bethlehem, Pa.
[7]Research metallurgist, Vanadium Corp. of America, Bridgeville, Pa.
[8]Metallurgist, Research Laboratory, General Electric Co., Schenectady, N. Y.
[9]Metallurgist, Research Laboratory, General Electric Co., Schenectady, N. Y.

number of us. There is a question as to whether or not this small grain might be due, in the first place, to this insoluble particle which was present in the austenitic state and formed the germ by which the grain was set up, or, as I believe Mr. McQuaid said he favored, the possibility of there being an inhibitor around the grains preventing them from growing. In some work which was presented here at the Society a year or so ago, on some strip steel, Mr. Brophy and I found that as we heated the strips to successively higher temperatures the decidedly normal steel gradually changed its size up to the critical and upon passing through the critical Ac there was not such a great change in grain-size.

In the abnormal steel, on the other hand, there was not much change in grain-size up to the critical, but upon passing the critical there was a very decided increase in grain-size.

If you will recall in Jeffries' and Archer's book, they cite the fact that is well known in incandescent lamp work, that ThO_2, when added to tungsten was put in there as a grain inhibitor. The solubility that takes place is nil. We do know that up to certain percentages of thoria, or up to certain conditions of treatment the inhibiting action is overcome, and we have the same thing that we found in the abnormal steel; that is, you can go from the smaller to the larger structure. This is, I believe, quite a good direct analogy.

Whether these particles are acting as nuclei for new grains, or whether they are inhibitors is open to question, and I think that this was the subject concerning which Mr. Foley wanted to commence a discussion.

Written Discussion: By W. E. Ruder, Research Laboratory, General Electric Co., Schenectady, N. Y.

Dr. Herty has again raised the question in this discussion as to the effect of silicon on the gases in molten steel. The idea that silicon merely takes care of the oxygen and so prevents its reaction with the carbon is plausible enough, but it is difficult to see how this can be the only explanation in such cases where steels of high alloy content, such as silicon, chromium, or aluminum, are remelted with from 4 to 10 per cent of the alloying element remaining in the bath. Such alloys are particularly difficult to keep in the mold, except by the addition of comparatively large amounts of aluminum. Such action would be expected if we assume that these elements, particularly silicon or aluminum, increase the solubility of the metal for gases, while the assumption that the silicon merely deoxidizes the iron seems hardly adequate.

The effect of deoxidizers on grain-size, as Mr. Foley and Mr. Wyman pointed out, is an important one. Grain-size is controlled by a number of factors, most important of which is the presence of "insolubles." These, however, must be in a state of division fine enough to actually cause lattice distortion before they can serve as nuclei for grains. We are quite familiar with the important effect of strain upon grain growth. With these finely divided particles acting as strain centers at innumerable points, crystallization of the metal starts and, depending upon the number of such starting points in a given area, the grain-size develops small or large by mutual limitation. It is quite possible for these "insolubles" to be intermetallic compounds as well as oxides. In any case, however, they must be small enough to affect, by their presence, the interatomic forces of the surrounding metal.

A. L. DAVIS:[10] Is it possible that the effect of aluminum additions to steel is in any way analogous to effect of iron additions to brass? The addition of 0.3 per cent of iron to brass reduces the grain-size resulting from a given anneal —the effect being most marked in the usual range of temperatures used for annealing, and less conspicuous if unusually high temperatures are employed. The very fine grain resulting in brass from addition of a small amount of iron results in a material increase in hardness with little or no loss of ductility.

Of course, we can see a dissimilarity due to the notable tendency of aluminum in steel to be mostly in oxidized form; but may there not be a small residuum of metallic aluminum which has its own specific effect in producing a finer grain-size?

Author's Closure

The several commentators on this paper have contributed much substantial information on the subject. Messrs. Greene and Luerssen bring up first the highly significant differences between the torsion and notched bar impact tests. An explanation of these differences must await development of an acceptable theory of failure under mechanical loading. In the interim the writer prefers to regard the notched bar impact test as applied to hardened steels simply as a criterion of grain-size.

Complete penetration of hardening in the impact test specimens is questioned. A section through a shallow hardening steel, No. 95, shows some fine pearlite, not over 10 per cent, near the center. The impact resistance of this steel nevertheless was consistently high. A significant difference in hardness between surface and bottom of notch is improbable except in untempered specimens. When tempered at 250 degrees Cent. or higher the hardness distribution is uniform since the differential tempering of quenching is eliminated.

Dr. Ruder states very clearly the case for the "mutual limitation" principle of grain growth inhibition. He also asks a question, the answer to which may be that reducing gases such as hydrogen can exist uncombined in "over deoxidized" steels, which would account for the wildness of which he speaks.

The point is made by Mr. Ashdown that certain open-hearth steels untreated with grain growth restraining agents were high in impact resistance. He implies that they are equal in all respects to the aluminum treated steels, but this is quite doubtful. The aluminum treated steels have a very high coarsening temperature, over 900 degrees Cent., and this is certainly the fundamental property with which we are most concerned here. It is highly improbable that the steels to which he refers have as high a coarsening temperature as the aluminum treated steels or that they are produced consistently without the aid of strong deoxidizers.

Aluminum treatment certainly increases materially the content of nonmetallic matter in the steel unless it is thoroughly deoxidized before the aluminum addition. When that is not done alumina streaks are quite prevalent and of course highly objectionable for certain types of service. Mr. Ashdown ascribes these streaks to the oxide film formed on solid aluminum when introduced in the molten steel, but such a skin would certainly float to the surface when it is broken and the enclosed molten aluminum released. These streaks

[10]Metallurgist, Scovill Mfg. Co., Waterbury, Conn.

are formed surely at the surface of the volume of steel containing a high concentration of aluminum, as this surface expands by diffusion of the aluminum from the point of introduction. On this basis the incidence of alumina streaks can be reduced by decrease in the available oxygen content of the bath and by introduction of the aluminum in less concentrated form and this statement is confirmed by experience.

Aluminum in high contents does increase the penetration of hardening when grain-size is held constant. The small content required for grain-growth inhibition, however, is insufficient to produce any notable effect. In fact penetration of hardening is greatly reduced, as Mr. Riegel states, with normal quenching practice and this is due to the finer grain size of the aluminum treated steel. Accordingly, when a coarse grain size has been a factor in obtaining the required penetration of hardening, steels having a high coarsening temperature should not be substituted without increase in alloy content to correct that effect.

Mr. Riegel's extensive observations, from which he concludes that notched bar impact resistance of hard steels correlates well with service failures, are noteworthy and supported by the writer's more limited observations. No contradictory views are expressed in the text of the paper.

Fig. 5(f), to which Mr. Riegel refers, is a very dirty steel picked originally for study with the expectation that any harmful effects of inclusions would be more prominent in hardened than in pearlitic steels. The textual remarks on purity of steel were inspired by observations on the melts made from hydrogen purified iron which are so free from nonmetallic matter that to find any requires extensive search. Our object in preparing these melts was to obtain high purity standards from which the influence of impurities and alloy content could be determined.

Mr. Dean's suggested explanation of the powerful effects of small aluminum additions is quite refreshing. Details of the experimental observations would be of great interest. Mr. Davis also suggests a solid solution effect of grain growth inhibiting additions of aluminum to steel from observations on the influence of iron additions on the recrystallization of brass. It is quite possible, however, that iron in brass reacts in the same manner as aluminum in steel. Mr. Wyman makes the interesting association of recrystallization of steel below the critical range with its normality.

In regard to Mr. Strauss' speculations, it may be said that oxygen and nitrogen can be reduced by hydrogen annealing to values within the limits of accuracy of the methods for their determination. Dr. Ruder's written discussion is pertinent to Mr. Foley's oral remarks. The observation of Mr. Brophy that dispersions of other materials than oxides restrain grain growth is quite pertinent. If it is true that the function of aluminum treatment is solely to produce nuclei for recrystallization, then carbides should perform the same function when properly dispersed. He implies, however, that there is a necessary relation between grain-size and abnormality which might properly be questioned.

TECHNICAL PROGRAM AND REPORTS OF OFFICERS AMERICAN SOCIETY FOR METALS—16th CONVENTION NEW YORK CITY, OCTOBER 1-5, 1934

FOR the benefit of those who were not in attendance at the Sixteenth Annual Convention of the Society and for purposes of record, the Technical Papers Program and the Reports of Officers read at the annual meeting are herewith published in full.

MONDAY, OCTOBER 1

Morning Session—Hotel Pennsylvania, Ball Room—10:00 A. M.

Joint Chairmen—D. F. McFarland and T. H. Wickenden

Heat Treating Session

Physical Properties of Case Hardened Steels, by O. W. McMullan, Timken Detroit Axle Co., Detroit.

Magnetic Properties of Iron as Affected by Carbon, Oxygen and Grain Size, by T. D. Yensen, Westinghouse Electric and Mfg. Co., and N. A. Ziegler, West Penn Electric Co., Pittsburgh.

Influence of Non-Metallic Inclusions Upon the Precipitation of Primary Cementite in Hypereutectoid Steels, by E. G. Mahin and E. F. Lee, University of Notre Dame, Notre Dame, Ind.

Afternoon Session—Port Authority Building—2:00 P. M.

Joint Chairmen—Norman I. Stotz and T. H. Nelson

Research Session

A Graphical Method for Calculating Heat Loss Through Furnace Walls, by C. E. Weinland, Johns-Manville Research Laboratories, Manville, New Jersey.

Properties of Some Cast Alloy Steels, by T. N. Armstrong, Norfolk Navy Yard, Norfolk, Va.

A Comparison of the Corrosion Resistance of Several Wrought Irons Made by Different Processes, by O. A. Knight and J. R. Benner, Pennsylvania State College, State College, Pa.

Practical Observations of Some High Carbon, High Chromium Tool Steels, by W. H. Wills, Ludlum Steel Co., Dunkirk, N. Y.

TUESDAY, OCTOBER 2

Morning Session—Hotel Pennsylvania, Ball Room—10:00 A. M.

Joint Chairmen—Kent R. Van Horn and R. F. Mehl.

X-Ray Session

Grain Distortion in Metals During Heat Treatment, by C. Nusbaum, Case School of Applied Science, Cleveland, and N. P. Goss, Cold Metal Process Co., Youngstown.

X-Ray Exposure Charts for Steel, by H. R. Isenburger, St. John X-Ray Service, Inc., Long Island City, N. Y.

New Development in Electrical Strip Steels Characterized by Fine Grained Structure Approaching the Properties of a Single Crystal, by N. P. Goss, Cold Metal Process Co., Youngstown.

Cutting Temperatures Developed by Single-Point Turning Tools, by O. W. Boston and W. W. Gilbert, University of Michigan, Ann Arbor.

Afternoon Session—Port Authority Building—2:00 P. M.

Joint Chairmen—John Johnston and A. E. White

Aging Session

Prolonged Tempering at 100 Degrees Cent., and Aging at Room Temperature of 0.8 Per Cent Carbon Steel, by G. A. Ellinger and R. L. Sanford, National Bureau of Standards, Washington, D. C.

The Effect of High Hydrostatic Pressures on Aging, by L. R. Van Wert, Lecturer on Metallurgy, Harvard University, Cambridge, Mass.

Directional Properties in Rolled and Annealed Low Carbon Steel, by Arthur Phillips and H. H. Dunkle, Yale University, New Haven.

WEDNESDAY, OCTOBER 3

Morning Session—Hotel Pennsylvania, Ball Room—10:00 A. M.

Annual Meeting of the A. S. M.

1934 Campbell Memorial Lecture, presented by Dr. V. N. Krivobok, Professor of Metallurgy, Carnegie Institute of Technology, Pittsburgh.

Chairman—Dr. Albert Sauveur

Afternoon Session—Port Authority Building—2:00 P. M.

Joint Chairmen—H. J. French and A. B. Kinzel

Research Session

Alloys of Iron and Manganese, Part XIII, by F. M. Walters, Jr. and Cyril Wells, Carnegie Institute of Technology, Pittsburgh.

The Influence of Oxide Films on the Wear of Steels, by S. J. Rosenberg and Louis Jordan, National Bureau of Standards, Washington, D. C.

The Metastability of Cemetite, by H. A. Schwartz, National Malleable and Steel Castings Co., Cleveland.

Heat Treatment in Controlled Atmospheres Including Gaseous Carburizing, by C. R. Austin, Penn State College, Penn State, Pennsylvania.

THURSDAY, OCTOBER 4

Morning Session—Hotel Pennsylvania, Ball Room—10:00 A. M.

Joint Chairmen—H. M. Boylston and N. L. Mochel

Research Session

The Cobalt-Molybdenum System, by W. P. Sykes and Howard F. Graff, General Electric Co., Cleveland.

Malleable Castings, by Albert Sauveur and H. L. Anthony, Harvard University, Cambridge.

Oxidation of a Low Carbon Steel in the Temperature Range 1650 to 2100 Degrees Fahr., by C. A. Siebert, U. S. Rubber Co., and Clair Upthegrove, University of Michigan, Ann Arbor.

The Ductility of Chromium-Nickel Austenitic Steels at Elevated Temperatures, by H. D. Newell, Babcock and Wilcox Tube Co., Beaver Falls, Pennsylvania.

Afternoon Session—Port Authority Building—2:00 P. M.

Joint Chairmen—C. H. Herty, Jr., and R. H. Harrington

Steel Making Session

Studies on the Metallurgy of Arc-Deposited Weld Metal, by S. L. Hoyt, A. O. Smith Corp., Milwaukee.

The Effect of Deoxidation on Some Properties of Steel, by C. H. Herty, Jr., Bethlehem Steel Co., Bethlehem, Pa.

Non-Metallic Inclusions in Steel, by S. F. Urban, National Tube Co., Ellwood City, Pa., and John Chipman, University of Michigan, Ann Arbor.

FRIDAY, OCTOBER 5

Morning Session—Hotel Pennsylvania, Ball Room—10:00 A. M.
Joint Chairmen—O. E. Harder and M. A. Grossmann

Grain Size Symposium

Grain Size in Metals with Special Reference to Grain Growth in Austenite, by M. A. Grossmann, Illinois Steel Co., Chicago.

General Relations Between Grain Size and Hardenability and the Normality of Steels, by E. S. Davenport and E. C. Bain, U. S. Steel Corp., Research Laboratories, Kearny, N. J.

Grain Size in Relation to Machinability and Other Properties of Bessemer Screw Steel, by H. W. Graham, Jones & Laughlin Steel Corp., Pittsburgh.

Grain Size Control of Open-Hearth Carbon Steels, by S. Epstein, Battelle Memorial Institute, Columbus, Ohio, J. H. Nead, Inland Steel Co., Indiana Harbor, Ind., and T. S. Washburn, Inland Steel Co.

The P-F Characteristic of Steel, by B. F. Shepherd, Ingersoll-Rand Co., Phillipsburg, N. J.

Effect of McQuaid-Ehn Grain Size on Hardness and Toughness of Automotive Steels, by H. W. McQuaid, Republic Steel Corp., Detroit.

Afternoon Session—Hotel Pennsylvania, Ball Room—2:00 P. M.
Joint Chairmen—E. C. Bain and Jerome Strauss.

Grain Size Symposium (continued)

Effects of McQuaid-Ehn Grain Size and Fracture Appearance as Influenced by the Nature of the Steel, by Philip Schane, Jr., Carnegie Steel Company, Pittsburgh.

Effect of McQuaid-Ehn Grain Size of Steel in Forging, by W. E. Sanders, Delco Products Division, General Motors Co., Dayton, Ohio.

Influence of Grain Size on the High Temperature Characteristics of Ferrous and Nonferrous Alloys, by A. E. White and C. L. Clark, University of Michigan, Ann Arbor.

The Importance of Grain Size of Sheet Steel for Deep Drawing, by R. L. Kenyon, American Rolling Mill Co., Middletown, Ohio.

The Influence of Grain Size on Magnetic Properties, by W. E. Ruder, General Electric Co., Schenectady, N. Y.

Factors Determining the Impact Resistance of Hardened Carbon Steels, by Howard Scott, Westinghouse Electric and Mfg. Co., East Pittsburgh.

ANNUAL ADDRESS OF THE PRESIDENT

Sixteenth Annual Convention, New York City, October 3, 1934

WILLIAM H. PHILLIPS, *President*

IT IS indeed a privilege to extend greetings to a representative gathering of the Society membership for the first time.

I would like, also, at this time to express our appreciation to the co-operating societies—The Institute of Metals Division and the Iron and Steel Division of the American Institute of Mining and Metallurgical Engineers, The American Welding Society, The Iron and Steel Division of the American Society of Mechanical Engineers and the Wire Association. It is through their continuing co-opera-

tion and assistance that we are able to make National Metal Congress week a large success.

It must be, to all of you who are closely associated with the local chapters or the National organization a source of pride to find again, at the close of another year, our society in such splendid condition. From the standpoint of membership, financial standing, educational work, chapter activity, we are surely forging ahead. But I will leave those phases of our activity to be reported on by our Secretary and Treasurer under whose jurisdiction they logically fall.

Many changes in our organization and its operation have taken place during the past year. They have been changes, however, that mark advance. We are operating under a new constitution. As you know, this became effective December 20th, 1933, and was the result of a very painstaking and comprehensive study of apparent faults in our old constitution. Some of us may have feared the wisdom of making some of the radical changes proposed. The results to date, however, have proven this wisdom.

One of the important changes involved the nomination and election of national officers. Through the co-operation of every chapter entering into the new arrangement with a determination to give it a fair trial, selected outstanding men as their potential representatives on the national nominating committee with the result that this committee, when finally appointed, quickly got together and unanimously selected a slate that can be depended upon to carry on the work of the Society. It has been pleasant to note the entire absence of political activity that formerly permeated the first day or two of our conventions.

In passing, I wish to express appreciation for the fine work of your first nominating committee under the new constitution. Chairmen, Prof. Bradley Stoughton, R. M. Pease, Arthur T. Clarage, G. R. Brophy, Gordon A. Webb, J. B. Johnson and F. P. Flagg.

The constitution now provides that officers newly elected will take office on day following our annual meeting. This means that present officers will serve for eight months and succeeding boards will serve from October to October.

Though our new name is but eight months old, it seems to have been very generally accepted and we more often hear Metal Society today than Steel Treaters.

Probably the greatest increase in society activity during the past year, has been in our educational work. By this we do not mean in

any sense are we becoming competitive with our schools of learning, rather are we stimulating interest that may carry on into the University. Much of the work we refer to is being supervised or actually carried on by members of a faculty of one of our institutions.

We have found, however, a real interest in a short lecture course of about six lectures. Where these courses have been given they have been well attended. For the most part this activity has been confined to the larger chapters, usually the admission fee to the course being membership in the society. The increase in membership of these larger chapters reflects the influence of the educational work. It was felt that while many of our smaller chapters had capable men to present these courses they were too few and it involved too much of their time to prepare such courses. However, if a course should be made available at National headquarters it would be rather extensively used. We now have such a course available entitled "Modern Metallurgy" prepared carefully by Dr. A. Allan Bates of Case School of Applied Science, Cleveland, and expect that during the coming year we will hear of its being used in a good many of the smaller chapters with as gratifying results as the larger chapters have experienced.

We have heard from time to time complaints that our technical sessions at the annual convention are becoming too theoretical. In the broadest sense our society brings together the practical man, the production metallurgist and the research metallurgist. To prepare a program that suits everybody is difficult. A great deal of highly theoretical work must be presented at this time. Your board has felt, however, that as an added educational feature it might be desirable to establish the practice of presenting a series of lectures by a well-known authority on some popular phase of metallurgy. If these lectures are well received one or more series will be arranged each year. This year J. P. Gill of the Vanadium-Alloys Steel Co., Latrobe, Pa., is giving a series of five lectures on the general subject of tool steel.

Due to repeated requests the board has authorized the indexing of the last ten volumes of TRANSACTIONS and it was made available February 1st of this year. It is a comprehensive and complete index and is being well received.

The membership of the society have shown a marked renewal of interest in sectional meetings this year. The tri-chapter meeting held in Columbus and the Pennsylvania Chapters' meeting held in

Penn State College, presented exceptional technical programs and were very well attended. The Penn State meeting was followd by the formation of the Penn State College Student group.

As you know, the society has been in the habit of presenting two awards each year, the Henry Marion Howe medal for the best paper presented before the Society during the year, and the Campbell Memorial lecture which is presented to a metallurgist whose past work insures his ability to present an outstanding lecture on some worthwhile metallurgical subject. We are justly proud of our Campbell Memorial lecturers and our Howe medalists and the splendid contributions which they have made.

Your board feels, however, there is another recognition we should make, and that to the individual who occasionally contributes an outstanding achievement marking metallurgical progress. There has accordingly been created a new medal to be known as the Albert Sauveur Achievement Medal. It is felt that this medal will not be presented every year, rather only when real achievement has been proven.

The first medal will be presented to Dr. Sauveur tomorrow evening.

It seems to me that the outstanding observation I have made during my encumbency of this office has been the splendid local organizations of our chapters and the interest these men have not only in each other and their local chapter but also in the national office. It explains why the American Society for Metals will continue to prosper.

In closing I wish to express my deep appreciation to the national board for their unanimous support and co-operation. I wish also to thank all of the committees for the interest and untiring effort they have put in their work and lastly, I wish to thank W. H. Eisenman, national secretary, and his efficient headquarters staff for their thoughtful interest and assistance not only to the officers of the Society but also to our entire membership.

REPORT OF THE TREASURER

October 3, 1934

EMIL GATHMANN, *Treasurer*

The unaudited balance sheet of the Society as of the close of business August 31, 1934, is as follows:

ASSETS			
Cash		$ 23,390.61	
Securities (at cost)		136,281.17	
Receivables less reserve		7,338.59	
Accounts with restricted banks		3,668.04	
Inventory		17,937.81	
Advances (for exposition, employees, chapters, post office, etc.)		6,075.55	
Office furniture and fixtures		3,600.00	
Prepaid expense (insurance, Metal Progress)		1,265.94	
1934 convention prepaid expense		13,571.83	
TOTAL ASSETS			$213,129.54
LIABILITIES			
Accounts payable (chapter refunds)		$ 399.73	
Reserves		45,100.00	
1934 convention advance receipts		23,198.95	
Surplus, Jan. 1, 1934	$139,331.63		
Plus net gain to Aug. 31	5,147.84		
Less surplus charges, 1933	48.61	144,430.86	
TOTAL LIABILITIES			$213,129.54

Our cash is approximately $9,000.00 more than a year ago, receivables $1,000.00 more and inventory about $4,000.00 less, while we still have $3,600.00 with restricted banks. The market value of our securities has continued to show improvement, the present value of $103,295.00 showing a gain of $1,677.00 or approximately 1½ per cent over the value of $101,617.50 reported last year.

Careful examination and consideration of our holdings by your treasurer and the board of trustees has not shown where our investments could be materially improved without considerable sacrifice and our investment account shows the same holding as of July 1, 1933, ($49,000.00 being in Government bonds).

We have not found it necessary to dispose of any securities to meet operating expenses and have again been able to operate the Society within its income and in fact add to our surplus.

As the fiscal year of the Society now starts with September 1, the books of the Society have been audited as of August 31, 1934, and show the same result as above. Considering that we have had a short period of only 8 months with no convention activity, upon which we usually secure a profit, the result is very gratifying.

Our income and expense statement as of August 31, 1934, covering the period from January 1, is as follows:

Income (eight months)	$77,448.81
Expense	72,300.97
Excess income over expense	$ 5,147.84

Income from Metal Progress has increased 30 per cent in the first eight months of this year over the same period in 1933, but expense has increased in approximately the same proportion and the net result is the same (about breaking even) although the volume

of business is greater. TRANSACTIONS and the REVIEW produce no revenue and must be added to our other fixed costs of general overhead expense.

The sales of books published, including the Stainless Steel Book and NATIONAL METALS HANDBOOK, has been better than expected and has helped materially in enabling the Society to show an addition to surplus.

The increase in membership dues' receipts, however, has been largely responsible for the fine position the Society shows in its financial statement ending August 31, 1934. There has been a steady increase in the membership, the gross receipts being approximately $7,000.00 more than last year.

It may be interesting to note that in addition to the assets of the National Society as shown by audit, the total cash assets of our chapters amounted to approximately $47,000.00 at the date of this report, which is an increase of about $7,000.00 over the report of last year.

The duties of your treasurer have been rendered most agreeable by the active co-operation of President Phillips, Mr. Eisenman, the board of trustees and Mr. Ohlson, of the Cleveland office.

CONDENSED AUDITED BALANCE SHEET

AMERICAN SOCIETY *for* METALS

As of August 31, 1934

ASSETS		
CASH		
On Hand	$ 4,308.53	
On Deposit	18,944.65	$ 23,253.18
SECURITIES		
U. S. Government Bonds	$ 49,831.90	
Other Bonds	83,775.00	
Accrued Interest	2,619.28	136,226.18
Accounts Receivable	$ 8,883.34	
Less—Reserve	1,497.39	7,385.95
Inventory		17,937.81
OTHER ASSETS		
Travel Advances, postage deposits, etc.	$ 6,208.16	
Deposits in Restricted Banks	3,668.04	9,876.20
Office Furniture and Fixtures less Depreciation		3,600.00
Prepaid Expense (Convention, Metal Progress, Insurance)		16,290.75
		$214,570.07
LIABILITIES		
Accounts Payable		$ 1,867.35
Reserves		45,100.00
Advance Convention Receipts		23,198.95
Surplus		144,403.77
		$214,570.07

The above from Balance Sheet prepared by Ernst & Ernst, Certified Public Accountants.

INCOME AND EXPENSE

AMERICAN SOCIETY *for* METALS

For the eight months ended August 31, 1934

INCOME			
"METAL PROGRESS"—MONTHLY PUBLICATION			
Advertising	$29,949.50		
Subscriptions	1,893.88		
Reprints	495.40		
Sales, binders, etc.	107.16	$32,445.94	
MEMBERSHIPS			
Membership dues	$49,024.20		
Sustaining exhibitors' dues	2,850.00	$51,874.20	
Less: Apportionment of dues to chapters	20,218.97	31,655.23	
Books published and purchased for resale		6,943.23	
Interest earned		3,866.32	
"Transactions"—bound copies, subscriptions, etc.		2,060.00	
Discount earned		376.16	
"The Review"—advertising and subscriptions		249.93	
Miscellaneous		6.29	
TOTAL INCOME			$77,603.10
EXPENSE			
AS SHOWN BY EXHIBIT			
"Metal Progress"—monthly publication	$34,960.70		
Books published and purchased for resale	5,446.44		
"Transactions"—bound copies, subscriptions, etc.	8,391.81		
"The Review"	1,641.58		
General expense	7,303.83		
Secretary's office	7,495.67		
Accounting department	2,975.11		
Directors' expense	1,120.80		
National committee's expense	1,037.23		
President's expense	570.26		
Recommended practice	360.50	$71,303.93	
Support of chapters		1,151.62	
Sundry merchandise purchased for resale		31.74	
TOTAL EXPENSE			72,487.29
NET PROFIT			$ 5,115.81

The above from Income and Expense sheet prepared by Ernst & Ernst, Certified Public Accountants.

ANNUAL REPORT OF THE SECRETARY

October 3, 1934

W. H. EISENMAN, *Secretary*

The American Society for Metals, on September 30, 1934, had a total membership of 5,443. Of this number 4,719 or 87 per cent were the member classification, 390 or 7 per cent were of the sustaining classification, 312 or 5.5 per cent were of the junior classification, with 22 honorary members.

On September 30, last year, the Society had a total membership of 4,468. This shows an increase for the year of 975, or 22 per cent.

This is undoubtedly one of the most outstanding records that it has been the Secretary's privilege to report to the membership. It shows continued and increased interest in the Society and may be

judged by all of the members of the metal industry in the work and accomplishments of the Society.

When one analyzes the accomplishments and services of the Society, there is immediately brought to mind the splendid work of the 40 chapters and groups which are the first line of defense of the Society and its most intimate contact with the individual membership.

Following that important detail, one may view with a pardonable degree of pride the various publications and other services which emanate from the national office.

Meetings of the Board of Trustees

Since the last report to the membership the board of trustees, has held 4 meetings, as follows:

Cleveland, January 18, 1934.
Cleveland, January 19, 1934.
Cleveland, August 3, 1934.
New York, October 2, 1934.

The reports of the first three meetings have been published in THE REVIEW, and the last meeting will also appear in a later number of that publication.

Educational Activities

Boston presented a 17-lecture course with an average attendance of 125. Chicago, 10 lectures, 132 average attendance. Cleveland presented two courses, one on welding having 125 attendance, and one on metallography having 160.

Detroit presented a 12-lecture course, average attendance 80.

Golden Gate presented a lecture-laboratory course in co-operation with the San Francisco public schools.

Los Angeles also had a lecture and laboratory course using their own laboratory equipment.

Hartford presented an educational course using the very unusual idea of spreading out for the convenience of those enrolling by giving the same lecture in five key cities of their territory.

Montreal presented a welding course of ten lectures, which were very well attended.

North West chapter was one of the first to conduct an educational course. In the year 1922, Dr. O. E. Harder, then of the University of Minnesota, conducted a course in metallurgy and heat treating.

Philadelphia, in addition to its usual co-operation with the metal-

lurgical course at Temple University, presented an 8-lecture course with splendid results.

York chapter co-operated with the Pennsylvania State College and conducted a class in metallurgy.

Plans for 1934 and 1935

The following chapters have indicated their intention of conducting educational courses during the present year (1934-1935):

Buffalo
Chicago—2 courses
Cleveland—2 courses
Detroit—2 courses
Golden Gate
Indianapolis
Lehigh Valley
Milwaukee
Montreal
New York
North West
Philadelphia
Pittsburgh
Rockford
Tri Cities
Washington
Worcester
York
Springfield

Howe Medal Award

The Henry Marion Howe Medal awarded annually for the paper of highest merit published in TRANSACTIONS during the year, was awarded for the year 1933 to Joseph V. Emmons, chief metallurgist of the Cleveland Twist Drill Company, Cleveland, Ohio.

For the year 1934 the committee of award has unanimously selected Mr. John Chipman, research engineer at the University of Michigan, for his paper entitled "Application of Thermodynamics to the Deoxidation of Liquid Steel," published in the May, 1934, issue of TRANSACTIONS.

Campbell Lecture

This lecture, established in 1925, by the board of trustees and honoring the memory of Edward DeMille Campbell of the University of Michigan, was presented in 1933 at the convention in Detroit, by Herbert J. French, metallurgist at the International Nickel Co. of New York. The lecture was entitled "Fatigue and the Hardening of Steels," and was published in full in the October, 1933, issue of TRANSACTIONS.

President's Medal

The President's Medal was presented to Past President Alexander H. d'Arcambal at the annual banquet during the Detroit convention in 1933.

The presentation and award were definite and tangible indica-

tions to Mr. d'Arcambal of the appreciation of the membership for his careful and enthusiastic administration of the affairs of the Society during his term of office.

Recommended Practice Committee

The personnel of the Recommended Practice Committee for the year 1934 is as follows: Dr. C. H. Herty, Jr., Chairman; H. J. Fischbeck, J. P. Gill, A. W. F. Green (resigned, Sept., 1934), W. D. Halsey, H. B. Knowlton, F. B. Lounsberry, E. C. Smith, and Jerome Strauss.

After the publication of the 1933 edition of the Handbook, the Recommended Practice Committee made a study to determine what new articles and sections should be included in future editions. A regular meeting of the Committee was held on March 12, 1934, in the National Offices of the Society and a working program formulated.

Technical Committee Assignments

As a result of the March meeting, the Recommended Practice Committee has at this time fifteen active sub-committees preparing articles or recommended practices on the following subjects:—

1. Aging of Iron and Steel.
2. Chromium Plating of Tools.
3. Forging of S. A. E. Alloy Steels.
4. Forging of Stainless Steels.
5. Forging of Tool Steels.
6. Gas Carburizing.
7. General Article on Heat Treatment.
8. Hardness Testing Methods.
9. Additional Hardness Conversion Data.
10. Heat Resisting Alloys.
11. Impact Testing.
12. Machinability of Steels.
13. Magnetic Testing.
14. Physical Constants of Metals.
15. Pyrometry.
16. Testing Methods for Tool Steels.
17. Wrought Iron.

Subjects assigned to Authors

The Recommended Practice Committee has also invited several authors to prepare articles on the following subjects:

1. Aluminum as an alloying element in heat resisting metals.
2. Aluminum as an alloying element and its relation to grain-size.
3. Aluminum as an alloying element in nitriding steels.
4. Blue brittleness.
5. Capacities, cost factors, etc. of oil, gas and electric furnaces.
6. Cold rolled bars and shafts.

7. Critical points for S. A. E. Steels.
8. Forging grades of carbon steels.
9. Hardness testing at elevated temperatures.
10. Inspecting, testing and physical properties of strip and sheet steel.
11. Constitution of Iron-Chromium-Nickel alloys.
12. Constitution of Iron-Cobalt Alloys.
13. Constitution of Iron-Manganese-Carbon alloys.
14. Constitution of Iron-Silicon alloys.
15. Constitution of Iron-Titanium alloys.
16. Metal cleaning by electrolytic methods.
17. Metal cleaning by shot and sand blast.
18. Metal cleaning by tumbling.
19. Metal cleaning by vapor degreasing.
20. Methods of chemical analysis.
21. Physical properties of large sections when made from both carbon and alloy steels.
22. Pickling.
23. The use of propane and butane as industrial fuels.
24. Slushing oils and greases.
25. X-ray radiography.

This makes a total of 42 articles which the Recommended Practice Committee has in process of preparation by sub-committees and authors.

As the Committee plans to consider the welding section of the Handbook at its next meeting, subjects on welding will probably be added to the program.

Nonferrous Subjects

The Nonferrous Data Sheet Committee of the Institute of Metals Division of the A. I. M. E. is having prepared, by authors, in co-operation with the Recommended Practice Committee, articles on the following subjects:

1. Bearing Metals.
2. Constitution of Aluminum-Nickel Alloys.
3. Etching of Aluminum for Metallographic Examination.
4. Polishing of Aluminum for Metallographic Examination.

Other articles will be prepared for the nonferrous section of the Handbook, but at this time the Committee have not completed their plans.

Handbook

Since the publication of the 1933 Edition (Sept.) of Metals Handbook, members and purchasers have been exercising their exchange privileges. The distribution of the Handbook is as follows:

Members exchange	3404
Issued to new members	1281
Sold to members of Institute of Metals Division of A. I. M. E.	500
Sold to non-members	665
Total Handbooks issued	5850

Therefore, a total of 4,685 HANDBOOKS have been exchanged by A. S. M. members or issued to new members. This leaves only 688 members of the Society who have not taken advantage of the Free Exchange Plan. This is a very gratifying response from the membership and exceeds the number of exchanges made, in the same period of time, for either of the previous editions.

Publication Committee

The Publication Committee for the year 1934 was made up of the following: Robert Sergeson, Chairman; Ray T. Bayless, Secretary; G. R. Brophy, C. L. Clark, Wayne Cockrell, A. E. Focke, T. L. Joseph, R. L. Kenyon, R. F. Mehl, C. B. Sadtler, W. P. Sykes, T. H. Nelson, M. R. Frommann, A. B. Kinzel and H. S. Taylor.

Throughout the year and up to the present date the committee has reviewed 54 papers, of which 41 have been approved, 9 rejected until after considerable revision was made and 4 definitely rejected and returned to the authors.

The first meeting for the year 1934 was held in Cleveland at the National Offices at 9:30 A. M. on June 1 and the following members were in attendance: Robert Sergeson, Chairman; Ray T. Bayless, Secretary; G. R. Brophy, C. L. Clark, Wayne Cockrell, A. E. Focke, T. L. Joseph, R. L. Kenyon, R. F. Mehl, C. B. Sadtler, W. P. Sykes and T. H. Nelson.

One of the suggestions worthy of mention which was offered was that reviewers of papers received between Convention periods, for publication in the TRANSACTIONS only, should be asked to offer suggestions to the editor as to possible reviewers or discussors for said papers in order that a comprehensive critique or discussion might be published along with the papers in TRANSACTIONS. Also it was suggested that the author of a paper be asked for suggestions as to the names of possible discussors of it.

The Publication Committee recommended to the board of trustees and it has been approved by them, that TRANSACTIONS be published as a quarterly rather than a monthly. The committee had in mind that this would permit a better grouping of subject matter and at the same time would effect certain economies which would be used as an improvement in the general make-up of the publication. As an example, the first issue of the quarterly, which will appear in December, will contain among other things all of the papers presented at the Grain-size Symposium scheduled for presentation at the 1934 New York Convention.

Transactions

There has been published since the last annual meeting the October, November and December, 1933, issues of TRANSACTIONS, the last three monthly issues of Volume 21, constituting 96 pages each or a total of 288 pages. There has also been published the first 9 issues of TRANSACTIONS for the year 1934, January, February, March, April, May, June, July, August and September, containing 96 pages each or 864 pages. The total pages published since the last annual meeting amount to 1,152.

The TRANSACTIONS published since October, 1933, contain papers presented at the 1933 Convention, together with others received during the year. The last three issues of Volume 21 contained 10 papers and the first 9 issues of Volume 22 for 1934, contained 34 papers. An analysis of the editorial production of TRANSACTIONS from Volume 1 up to and including Volume 22, No. 9 (September, 1934) shows a grand total of 20,999 pages.

In February of this year the Society published a general index to TRANSACTIONS, Volumes 11 through 20 and also a general index to METAL PROGRESS for 1930 through 1932.

This index was made available to the membership of the Society in two forms, first a bound book containing a general index to Volumes 11 through 20 of TRANSACTIONS containing 116 pages and METAL PROGRESS, consisting of 161 pages; and second a bound volume containing 216 pages covering the index to Volumes 1 through 20 of TRANSACTIONS and METAL PROGRESS.

The index to TRANSACTIONS was compiled by Frank T. Sisco and the index to METAL PROGRESS was compiled by Marjorie M. Rud.

Preprints

30 papers were presented before the 15th annual Convention held in Detroit last year (1933) 22 of which were preprinted. For this year's Convention 37 papers have been scheduled for presentation, 31 of which have been preprinted and distributed to those members requesting them. The total number of pages of preprints for 1933 was 452, whereas the total number of preprint pages for 1934 is 620. The number of requests for preprints this year has been approximately double that of last year.

Review

During the past year THE REVIEW has been published as a bimonthly containing on the average, 6 pages per issue.

The September issue, which had a circulation of 20,000 among the members of all societies and divisions co-operating in the 1934 National Metal Congress, was 8 pages.

From November, 1933, to date (6 issues) 1,633 column inches were devoted to editorial matter and 730 to advertising. The amount of THE REVIEW space devoted to reports of chapter meetings was 909 column inches, or slightly more than one-half of the entire space during the year.

Metal Progress

METAL PROGRESS has now been coming to you every month for four years, and needs no words of mine to commend it. The editor has kept constantly in mind the wishes of the Board of Trustees that the magazine should supplement the more technical work of the TRANSACTIONS, with briefer articles about the production and application of interesting new metals and alloys, new shop practices and equipment, and thus interest the members whose daily work is with the production, sale and fabrication of metal rather than its scientific study. We feel that METAL PROGRESS, in following this editorial objective, has been an important item in the services to the members.

The financial success of the special October issue, now in your hands, encourages us very much. Every extra advertisement in that issue was placed without high pressure salesmanship, and is there because the members of the American Society for Metals represent a very important group of consumers of the material or equipment advertised and because METAL PROGRESS is unquestionably the best and most economical medium wherewith to reach that audience.

As the American Society for Metals closes its short year of eight months, the trustees and the members may look in retrospection and observe a number of outstanding indications of healthy growth and a bigger and better outlook for the future.

With all of the Society activities showing an income of over $5000.00 in excess of its expenses, with the present issue of METAL PROGRESS carrying the largest volume of advertising since it was established four years ago, with the membership of the Society approximately 1000 members more than a year ago, with all advertising income 30 per pent increased over the previous period, and with the National Metal Exposition the largest since 1929, and showing a 45 per cent increase over the amount of space sold for the

exposition held in Detroit last year, with an increased educational activity on the part of the chapters, and with the continued growth of the services from the chapters and the national office, the membership of the Society may look to the future with a degree of confidence that the present success of the Society will be maintained.

NOMINATION OF OFFICERS

W. H. PHILLIPS, *President*

I have the pleasure to report that in accordance with the constitution of the A. S. M. the President received prior to March 1 of this year suggested candidates for the Nominating Committee from all the local chapters of the Society. From the list of eligible candidates suggested, I appointed the following Nominating Committee of seven:

Professor Bradley Stoughton of Lehigh Valley as chairman.
A. T. Clarage, Chicago.
Robert M. Pease, St. Louis.
J. B. Johnson, Dayton.
Gordon A. Webb, Detroit.
Gerald R. Brophy, Schenectady.
F. P. Flagg, Boston.

The names of the members of the Nominating Committee were published in the March REVIEW.

The Nominating Committee responded to the call of the president and held a meeting in New York City during the third full week in May and reported immediately the selection of the following candidates for the offices indicated:

For president for one year—B. F. Shepherd.
For vice president for one year—R. S. Archer.
For secretary for two years—W. H. Eisenman.
For trustee for two years—Walter Mathesius.
For trustee for two years—George B. Waterhouse.

The report of the Nominating Committee was published in the May REVIEW.

I have been notified by the secretary that no additional nominations were received prior to July 15 for any of the vacancies occurring on the board of trustees. Consequently, the nominations were closed. I shall now call upon the Secretary to perform his duty as prescribed by the constitution.

W. H. EISENMAN, *Secretary:* Mr. President, since no additional nominations were received prior to July 15, 1934, for any of the vacancies on the board of trustees of the American Society for

Metals, I herewith, in compliance with the provision of the constitution, cast the unanimous vote of all of the members of the Society for the election of the following candidates:

For president—for one year—B. F. Shepherd.
For vice-president—for one year—R. S. Archer.
For secretary—for two years—W. H. Eisenman.
For trustee—for two years—Walter Mathesius.
For trustee—for two years—George B. Waterhouse.

W. H. PHILLIPS, *President:* Since all the provisions of the constitution have now been performed, I herewith declare the candidates named to have been duly and unanimously elected for the term and office specified, and the term of office of each officer just elected shall begin on the day following the close of this annual meeting.

President's Bell

"At this time we usually make the presentation of the President's Bell to the chapter who, in the opinion of the President, has performed from all standpoints the best during the past year. We are going to do something this year that has not been done before. Usually the President's Bell has been passed around, but after making a very careful study of the activities we found about three chapters that were running neck and neck, but we have finally decided that the Cleveland chapter's all-around performance for the year 1934 entitled them to retain the Bell for another year.

"It gives me great pleasure to again present to you, Dr. Kent R. Van Horn, as a representative of the Cleveland chapter, the President's Bell for another year."

Annual Banquet

In addressing those in attendance at the Sixteenth Annual Banquet of the society, President Phillips said in part:

"Members of the American Society for Metals and honored guests:

"We welcome you to this our Sixteenth Annual Banquet. It is our first opportunity to meet in New York but hospitality shown us by the City fathers, the convention committee, our own New York and New Jersey chapters, and hotels alike insures our return at an early date.

"I would like to express the appreciation of the officers and membership of The American Society for Metals, The Institute of Metals and the Iron and Steel Division of the American Institute of

Mining and Metallurgical Engineers, The Iron and Steel Division of The American Society of Mechanical Engineers, The American Welding Society and the Wire Association, for their hearty co-operation, which makes possible National Metal Week."

At this point the guests at the speaker's table were individually introduced to the audience.

Presentation of Past President's Medal

In continuing President Phillips said: "It now becomes my privilege and duty to present the Past President's Medal to Mr. W. B. Coleman. For years Bill has worked tirelessly in the interest of the society—first in the Philadelphia chapter, where he held all offices—later in the National Society where he did conspicuous work as the Chairman of our meetings and papers committee. Having sat on the board during his tenure as President, I am well acquainted with the energy and painstaking care he has exercised in all society matters. He was able to personally visit 90 per cent of our 39 chapters and groups, and worked very hard to secure a better understanding and co-operation on the part of industry with our Society. Bill—on behalf of the officers and membership of the American Society for Metals—may I present to you the Past President's Medal."

Henry Marion Howe Medal Award

"Every year, some time prior to this banquet, we start to select the outstanding paper presented before the Society and published in TRANSACTIONS during the year. Four committees are appointed consisting of three men each. These committees each cover three month's publications, selecting the two best papers of that quarter. The final award committee, of three members, select from the eight papers thus chosen one paper, the author of which is presented the Henry Marion Howe medal. This year we are pleased to present to Mr. John Chipman, of the University of Michigan, the Henry Marion Howe Medal for his paper entitled, "Application of Thermodynamics to the Deoxidation of Liquid Steel," presented at last year's convention and published in the May, 1934, issue of TRANSACTIONS."

Albert Sauveur Achievement Medal Award

"As explained in my annual report yesterday, we, as a Society, are justly proud of our Campbell Memorial Lecturers and our Howe Medalists and the contributions which they have made to our metal-

lurgical literature. We have felt, however, an outstanding achievement in the field of metallurgy should receive special recognition—so after a great deal of deliberation, we have created a new medal to be known as the Albert Sauveur Achievement medal. This award is in the name of and in honor to one who has contributed so richly and inspired so universally workers of all classes in the field of metallurgy. It is highly appropriate, therefore, that the American Society for Metals should designate its Achievement Medal as the Albert Sauveur Achievement Medal in honor of one who is, beyond question, the peer of all metallographists. We wish to make the first presentation of this medal this evening. I am going to ask Dr. A. E. White, the first president of our society, to present to us the first candidate for this honor.

In presenting Dr. Sauveur, Dr. White commented as follows:

"As the first recipient of the Albert Sauveur Achievement Award, I am pleased to announce that your Board of Trustees has selected our esteemed friend and co-worker Dr. Albert Sauveur, Gordon McKay Professor of Metallurgy at Harvard University.

"This award is made because of the great contributions Dr. Sauveur has made in adding to our knowledge of the structural changes which take place in iron, steel and other metals due to heat treatment. He has pioneered in this field. With a simple microscope mounted on a barrel, he began his work. His research equipment has always been limited, but in spite of or because of this condition, he has shown to the world the possibility of achievement in one's chosen field when one approaches his work with boundless enthusiasm.

"Dr. Sauveur was born June 21, 1863, at Louvain, Belgium, of French parentage. He was educated at Athenee Royal, Brussels, School of Mines, Liege, and Massachusetts Institute of Technology. He served as Chemist and Metallurgist first for the Pennsylvania Steel Company and then for the Illinois Steel Company from 1889 to 1897. He has been at Harvard University since 1899.

"During the World War he served as Metallurgist for the American Aviation Commission from 1917 to 1919, and Metallurgical Expert to the French Ministry of Munitions, and later as Director of Research for the Metallurgical Division of the American Expeditionary Force.

"From 1898 to 1903 he was Editor of *The Metallographist*, and from 1903 to 1906 Editor of the *Iron and Steel Magazine*. He has published many publications in the field of metallurgy especially on

such subjects as the constitution of steel. His outstanding book is The Metallography of Iron and Steel. This is used as a standard text by many of the schools and colleges in this country.

"Dr. Sauveur is a Fellow of the American Academy of Arts and Sciences, an honorary member of the Society of Engineers, and honorary member of the American Society for Metals, as well as an honorary member of other international organizations.

"He has had many honors conferred upon him as a result of his achievements. He is an Officier d'Academie, Chevalier Legion d'Honneur and President, Salon Francais, Boston, 1920-1924. He served as United States Delegate to the Third Pan American Scientific Congress at Lima, Peru, 1924. He was awarded the Elliot Cresson gold medal by Franklin Institute in 1924, awarded the Bessemer Medal by the British Iron and Steel Institute in 1924 as well as other honors.

"His work has also been recognized by various Universities, having been given a degree of Doctor of Science by Case School of Applied Science, by the University of Grenoble in 1924, and the University of San Marcos in 1925. In 1926 he was given the honorary degree of Doctor of Engineering by Lehigh University.

"On behalf of the Board of Trustees of the American Society for Metals, it is my very great honor, Mr. President, to present to you Dr. Albert Sauveur as the first recipient of the Albert Sauveur Achievement Award of the American Society for Metals."

In presenting the award to Dr. Sauveur, President Phillips commented as follows:

"Dr. Sauveur—It indeed gives me great pleasure to perform the duty which falls to me and on behalf of the American Society for Metals, I present you with the first Albert Sauveur Achievement Medal."

In accepting the award Dr. Sauveur responded as follows:

"Mr. President, my friends, ladies and gentlemen:

"I find it difficult to express adequately my appreciation of the unusual honor which is being conferred upon me today.

"That you should have decided to do so while I am still with you adds a great deal to its significance!

"In spite of what has been said tonight, I fully realize that I owe this honor to your indulgence and generosity more than to what I may have contributed to the art and science in which we are so deeply interested.

"May the award of this medal in years to come stimulate younger

men in their efforts to advance the ancient and basically important metallurgical art! I find in the possible realization of that hope the richest reward for what it has been my good fortune to be able to do in those early days when metallography was in its infancy and found it so difficult to grow to manhood. It is now so sturdy that all efforts to undermine its health have been abandoned. I have no desire to bore you with reminiscences. I think too much of you for that. I should like, however, to illustrate very briefly some of the difficulties encountered in those early days.

"I started my metallographic work in 1891 at the suggestion and under the inspiration of W. R. Walker at the time General Manager of the Illinois Steel Company. After five happy years of work in a field but recently opened, a hurricane in the form of a new president, John W. Gates by name, struck the Illinois Steel Company, which in its violence carried away the metallographical laboratory and its occupants. Being then settled in Chicago and with my first born but a few weeks old, I made some effort to remain in that city. I attempted to convince President Harper of the newly established University of Chicago that metallography should be taught at his institution and that I was the man to do it. In this I failed.

"We went back east and my next effort was to interest the Carnegie Steel Company to take up metallography. A young man by the name of Charley Schwab was the superintendent and after consulting his colleagues, among whom were Corey and Unger, he informed me that he had nothing to offer. My next attack was directed at the Watertown Arsenal—going to Washington to interview General Flager, at the time Chief of Ordnance. After several weeks delay, I received a long letter in which he expressed the fear that if the work was undertaken at the Arsenal it might prove so fascinating that Mr. Howard, who was then in charge of the Testing Laboratory, might lose his interest in everything else.

"I am vindictive enough to wish that President Harper, John W. Gates, Charley Schwab and General Flager might be here tonight —and that I could guide them through the metal exposition where they would find the walls covered with photomicrographs of steel and other metals and place before them the many papers presented at this meeting and so profusely illustrated with similar photographs.

"With all these rebuffs before me, I decided to open some laboratories in Boston. I even had the temerity to undertake single-handed and without any financial backing the publication of a quarterly maga-

zine known as *The Metallographist*. I have often wondered at this audacity of youth. Had I been older or wiser, I should never have dared to go into such a venture. I believe, however, that it did serve a useful purpose.

"I should like to be permitted to express one thought pertaining to the progress of metallurgy in future years. I find it in the parable of the light and the bushel. How many when they believe to have discovered a little light proceed immediately to hide it under a bushel so as to keep it for their exclusive use and in the hope that their neighbors may be left to stumble in the dark.

"Closer examination often reveals that the hidden light is a feeble and flickering one, hardly worth the cost of the bushel. At other times, the bushel is so poorly woven that it permits the leaking of the light, while in other instances still the light is so brilliant that no bushel can shield it. And all the while each one is trying to peep under the bushel of his neighbors. We may liken this situation to a large field covered with bushels assumed to shelter lights of various intensities, including those of no intensity at all.

"How much better it would be, if all those bushels could be lifted and discarded, leaving a brilliantly illuminated field—to the benefit, I believe, of the metallurgical industry considered collectively or individually.

"While I am not prepared to relinquish this beautiful medal—indeed anyone who would try to take it away from me would have a hard fight on his hands—I think in all frankness that the first award should have been made to the American Society for Metals itself. If it is to be an award for achievements, who more than this Society deserves it!

"Indeed, its achievements during so short an existence have been a source of wonder and admiration on the part of other older scientific and technical societies.

"I accept this medal then, not as my due, but as an evidence of the fine sentiments that prompted you in establishing it. It testifies of the earnest desire of this society to show its appreciation of the unselfish labor of those engaged in serving mankind in an important field—Long live the American Society for Metals!"

Following this presentation the speaker of the evening, Mr. T. M. Girdler, chairman of the board of the Republic Steel Corp., gave his outstanding address, "Tomorrow's Market for Steel."

Statement of the ownership, management, circulation, etc., required by the act of congress of August 24, 1912, of TRANSACTIONS of the American Society for Metals, published quarterly at Cleveland, Ohio, for October 1, 1934, State of Ohio, county of Cuyahoga, ss. Before me, a notary public, in and for the state and county aforesaid, personally appeared Ray T. Bayless, who, having been duly sworn according to law, deposes and says that he is the editor of the TRANSACTIONS of the American Society for Metals, and that the following is, to the best of his knowledge and belief, a true statement of the ownership, management, etc., of the aforesaid publication for the date shown in the above caption, required by the Act of August 24, 1912, embodied in section 411, Postal Laws and Regulations to wit:

1.—That the names and addresses of the publisher, editor, managing editor, and the business managers are: Publisher, American Society for Metals, 7016 Euclid Avenue, Cleveland, Ohio; Editor and Managing Editor, Ray T. Bayless, 7016 Euclid Avenue, Cleveland, Ohio; Business Manager, W. H. Eisenman, 7016 Euclid Avenue, Cleveland, Ohio.

2.—That the owner is: The American Society for Metals, 7016 Euclid Avenue, Cleveland, Ohio, which is an educational institution, the officers being: President, B. F. Shepherd; Vice-president, R. S. Archer; Secretary, W. H. Eisenman; Treasurer, Emil Gathmann; Directors, W. H. Phillips, E. C. Bain, W. P. Woodside, Walter Mathesius, George B. Waterhouse. All above officers of the American Society for Metals, address at 7016 Euclid Avenue, Cleveland, Ohio.

3.—That known bondholders, mortgagees and other security holders are: none.

4.—That the two paragraphs next above, giving the names of the owners, stockholders and security holders, if any, contain not only the list of stockholders and security holders as they appear upon the books of the company but also, in cases where the stockholder or security holder appears upon the books of the company as trustee or in any other fiduciary relation, the name of the person or corporation for whom such trustee is acting, is given; also that the said two paragraphs contain statements embracing affiant's full knowledge and belief as to the circumstances and conditions under which stockholders and security holders who do not appear upon the books of the company as trustees, hold stock and securities in a capacity other than that of a bona fide owner; and this affiant has no reason to believe that any other person, association, or corporation has any interest direct or indirect in the said stock, bonds, or other securities than as so stated by him. Ray T. Bayless, managing editor, sworn to and subscribed before me this 1st day of October, 1934.

(Seal) Arthur T. Wehrle, notary public.

(My commission expires January, 1935.)

"TOOL STEELS"

A Series of Five Educational Lectures on the Selection, Properties and Uses of Commercial Tool Steels

by JAMES P. GILL

Chief Metallurgist

VANADIUM-ALLOYS STEEL CO., LATROBE, PENNA.

This five-lecture series constituting a most outstanding piece of work by an expert in the tool steel industry, will make a valuable addition to your library. These lectures were presented at the Sixteenth Annual Meeting of the Society in New York City, October 1 to 5, 1934, and are now made available to the entire membership. This is a book containing 136 pages and is neatly bound in cloth. It is available in paper cover also.

Price: Cloth bound, $2.50—Paper bound, $2.00

CONTENTS:

Lecture I. Manufacture, Classification and Selection of Tool Steels
- a. History
- b. Manufacture
  Melting, forging, annealing and inspection
- c. Classification
- d. Selection
  Correlation of physical properties and requirements of applications

Lecture II. Evaluation and Testing: Effect of Alloying Elements
- a. Evaluation
  Shop test, quality and physical properties
- b. Testing
  Tensile properties, hardenability, fracture, macro and microscopic tests
- c. General effect of Alloying Elements
  C, S, P, Mn, Si, Ni, Cr, V, Cr-V, W, Mo
- d. Specific Effect of Alloying Elements on Hardenability

Lecture III. Carbon and Low Alloy Tool Steels
- a. Straight Carbon Tool Steels
- b. Carbon Vanadium Tool Steels
- c. Carbon Tool Steels with Less than 1.00% of Alloying Element
- d. High Carbon, Low Tungsten Tool Steels
- e. Manganese Oil Hardening Tool Steels

Lecture IV. Tool Steels of Intermediate Alloy Content
- a. Low Carbon, Low Tungsten Types
- b. Finishing Steels
- c. Silicon-Manganese Steels
- d. Die Steels for Hot Work
  Chrome die steels, tungsten steels and chrome-tungsten steels for hot work, miscellaneous types

Lecture V. High Carbon, High Chromium Steels and High Speed Steels
- a. High Carbon, High Chromium Steels
  Air hardening types, oil hardening types
- b. High Speed Steels
  18-4-1 types, 18-4-2 types, 14-4-2 types, cobalt types, miscellaneous types

Send your check or money order to the

American Society for Metals

7016 Euclid Avenue — Cleveland, Ohio